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PROCEDURES FOR AVOIDING HEAT-TREAT CRACKING IN NICKEL-BASE SUPERALLOY WELDMENTS

A. T. D'Annessa

Lockheed-Georgia Company

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A. T. D'Annessa

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FOREWORD

This report presents results of a study of "procedures for avoiding post-weld heat-treat cracking in nickel-base superalloy weldments" under Contract F33615-71-C-1135, Project No. 7351, "Behavior of Metals", Task No. 735102, "Welding and Brazing of Metals and Components for Military Aerospace Structures and Jet Engines", conducted by the Lockheed-Georgia Company, Marietta, Georgia, for the Air Force Materials Laboratory, AFSC, Wright-Patterson Air Force Base, Ohio, with Dr. Guinn E. Metzger (LLP) serving as Project Engineer. The report describes results of research conducted during the period 4 January 1971 through 31 December 1971 in the Lockheed-Georgia General Structures and Materials Laboratory, with A. T. D'Annessa serving as the Principal Investigator.

Acknowledgements are extended to J. S. Owens, formerly with the Lockheed-Georgia Company, for his assistance during the first six months of the program, and especially for his contributions in the design and application of the acoustic emission system used throughout the contract and to A. D. Friday for his excellent light and electron microscopy work.

This technical report has been reviewed and is approved.

I. Perlmutter
Chief, Metals Branch
Metals and Ceramics Division
Air Force Materials Laboratory

ABSTRACT

This is a report of a program concerned with the development of procedures for avoiding post-weld heat-treat cracking in nickel-base superalloy weldments. Crack-susceptibility test procedures and an acoustic emission technique developed in a prior program were used in this program. Crack-susceptibility data were obtained for a number of studies including the effects of varying pre-weld solutioning treatments, effects of pre-weld cold work, effects of oxidation damage due to inadequate shielding during welding, evaluation of Rene' 41 powder metallurgy sheet material, procedure verifications, and the evaluation of varying pre-weld base metal heat treatments for avoiding heat-treat cracking in Astroloy. The 1975°F-1/2 hr (WQ) solutioning treatment was found to be an optimized pre-weld heat treatment for avoiding post-weld cracking in Rene' 41. Post solution quench rate was found to be a critical variable with slower quenches increasing the susceptibility of Rene' 41 to heat-treat cracking. Small amounts (to 2%) of pre-weld cold work did not reduce the effectiveness of pre-weld solutioning treatments for avoiding heat-treat cracking. The 1975°F-1/2 hr (WQ) pre-weld solution treatment was found to be effective in avoiding post-weld heat treat cracking in three heats of Rene' 41 and one heat of Waspaloy. Promising pre-weld overaging treatments for Astroloy were found to be detrimental in that they markedly increased the material's susceptibility to cracking during welding. The report also covers on-heating hardening response and detailed microstructure studies associated with varying pre-weld base metal conditions.

For internal control purposes, this report has been assigned Lockheed-Georgia Company Report Number ER-11272.

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SECTION I
INTRODUCTION

The work reported here is concerned with the development of procedures for eliminating or minimizing the persistent problem of post-weld heat-treat cracking in precipitation-hardenable nickel-base superalloy weldments. This cracking is found in the heat-affected zone of fusion welds and is often referred to as "strain-age cracking". It occurs during post-weld heat treatment generally during on-heating within the aging temperature range. The heat-treat cracking problem is of continuing concern with regard to component costs and design optimization. Less-than-optimum design approaches are frequently used to avoid existing and potential difficulties with complex weldments resulting in unnecessary weight penalties. Also, costs due to repair-reclamation and scrap loss of welded superalloy parts have not declined appreciably. In addition, the problem is a major deterrent in the design and fabrication of higher temperature and high-performance engines. Major advances in engine technology and other high temperature applications would be possible by simply improving the fabricability of existing high hardener (Al + Ti) content alloys such as Astroloy (Udimet 700). Consequently, there is an urgent need for solving the problem of heat-treat cracking in both the presently used intermediate hardener content superalloys and the less frequently used higher hardner content alloys which are more prone to post-weld heat-treat cracking.

Scope of Program

The objective of this program was to develop procedures for avoiding heat-treat cracking in the currently used alloys such as Rene' 41 and Waspaloy and then to evaluate the suitability of similar procedures for avoiding cracking in the more susceptible alloys such as Astroloy. The phases of work reported here are, for the most part, extensions of phases of work conducted in a preceding AFML-sponsored contract[1]. Major efforts of this program included:

1. Further evaluation of pre-weld heat treatments for Rene' 41 and the applicability of these heat treatments for avoiding cracking in other heats of Rene' 41 and Waspaloy.

2. Continuation of studies of pre-weld heat treatments for minimizing the incidence of cracking in Astroloy.

3. Determine the specific effects of heat treat atmosphere on post-weld cracking.

4. Complete aspects of several studies initiated in the preceding program[1] including the evaluation of fine grain powder metallurgy sheet material and the effects of small amounts of pre-weld cold-work on cracking behavior.

SECTION II

RESULTS OF PRIOR WORK

The primary purpose of this section is to review the results obtained in the preceding contract effort[1] which constitutes much of the background for the work described here. Prior pertinent work by other investigators, reviewed in Reference 1, may be found in References 2 through 13; more recent work is covered in References 14 and 15. The limited success during the past decade in finding suitable solutions to the problem of heat-treat cracking and the frequent inability to correlate results of work from differing sources may be attributed to the following reasons:

(1) The number of known and, possibly, unknown variables influencing cracking behavior and the complexity of their interactions.

(2) Inability to isolate each variable systematically in order to determine its specific influence on cracking.

(3) Lack of a reliable and reproducible test capable of discriminating the heat-treat cracking problem and variations in cracking behavior due to test parameter variations.

(4) Inability to obtain quantitative information from crack-susceptibility tests with respect to the temperature dependence and nature of cracking.

Most of these difficulties were overcome in the preceding contract effort with the development of a reliably, reproducible, new circular-patch crack-susceptibility specimen and a technique for acoustic emission monitoring of heat-treat cracking events. These are described in detail in subsequent sections of this report.

Effects of Variables on Heat-Treat Cracking

Base Metal Composition

Since commercial-grade heats of material were used in the preceding program, there were no correlations made of the effects of minor alloy elements such as

iron, manganese, sulfur, and silicon on heat-treat cracking behavior. It was felt that commercial melting practice was adequate for keeping these residuals under control and, therefore, minor element variations were not considered variables. A low-carbon heat of Rene' 41 was evaluated and found to be more crack-susceptible than the higher-carbon commercial grades; this observation was contrary to that reported in earlier research[9] and consistent with the results noted in Reference 10. This would indicate that the effects of carbides and carbide morphology, within the limits of carbon contents of commercial grades, are not a significant factor in heat-treat cracking. Consequently, it is apparent that cracking behavior is primarily dependent on hardener (Al + Ti) content and γ' characteristics.

Welding Process Parameters

Crack-susceptibility data were obtained screening the effects of weld energy input, simulated repair welding, and tack welding. As was also concluded by other investigators, a lower weld energy input increased the resistance to heat-treat cracking which may be attributed to lowering the residual weld stresses rather than altering the heat-affected zone microstructure. The influence of simulated repair welding and inadequately shielded tack welds on heat-treat cracking was found to be negligible. It was apparent from these results that heat-treat cracking could not be avoided by the manipulation of weld process parameters alone. However, lower weld energy input techniques would appear desirable for use with any general control plan concerned with reducing the incidence of heat-treat cracking.

Microstructure

Several "bulk" microstructure considerations were studied in the preceding contract effort. These included the effects of base metal grain directions and grain size and a comparison of the resistance to heat-treat cracking of powder metallurgy product sheet and conventional wrought product material. The effects of base metal grain direction were clearly demonstrated by crack-susceptibility test results. The characteristic cracking pattern was found to shift with changes in the orientation of the grain direction of the specimen disk and secondary radial cracking was almost always influenced by grain direction. Results with Rene' 41 powder metallurgy sheet

material converted from extrusion-consolidated REP (Rotating Electrode Process) bar stock indicated that its resistance to crack-initiation was comparable to that of conventional wrought Rene' 41 material; however, the resistance to gross cracking was considerably better than that of conventional wrought product. Further isothermal crack-susceptibility evaluations, reported in the present work, were considered necessary to establish a more complete correlation of the resistance to heat-treat cracking of the REP and conventional sheet materials. The influence of grain size was not clearly established due to the inability to separate effects due to chemical composition and pre-weld processing history which affect γ' precipitation kinetics.

Pre-Weld Base Metal Processing

A number of pre-weld base-metal thermal and thermal-mechanical treatments was studied as an approach to "desensitizing" the heat-affected microstructure. These included several precipitate conditioning, solutioning, and solution-cold working treatments. The use of a 1975°F-25 minute (WQ) solutioning treatment was found to be effective in avoiding on-heating cracking in Rene' 41 and increasing the gross cracking temperature of Astroloy by 100° to 125°F. This treatment was also effective for avoiding on-heating cracking in a crack-sensitive low-carbon heat of Rene' 41. Crack-susceptibility data showed that pre-weld precipitate conditioning and intermediate cold working also affected on-heating cracking characteristics. However, the improvements in the resistance to heat-treat cracking observed with these pre-weld conditions were not as significant as that noted with the simple pre-weld solutioning treatment. The pre-weld solutioning treatment was also found to be as effective, if not more so, than a previously developed overaging treatment for avoiding heat-treat cracking in Rene' 41.

Heat-Treat Atmosphere

Crack-susceptibility data obtained with various heat-treat environments revealed that furnace atmosphere influenced the cracking behavior of Rene' 41. A hydrogen environment was found to have the most pronounced effect as revealed by an increase in gross cracking temperatures. Results with partial vacuum environments were considered questionable due to several aspects of the test specimen design and

the method used for evacuation. The most notable effect, revealed by acoustic emission analyses, was a shift from rapid (unstable) crack extension to slow (stable) crack growth with the hydrogen, helium, and partial vacuum environment tests. This shift was interpreted as an increase in the resistance to heat-treat cracking. Results from these preliminary tests were considered inconclusive, but sufficiently encouraging to warrant further evaluation in the present program. The data from the present effort, contained in a later section, compares the heat-treat cracking characteristics of tests conducted in air versus a vacuum environment using an improved crack-susceptibility specimen. These tests were considered necessary to clarify conclusions of other investigators[13] using a somewhat different approach in assessing the effects of furnace environment on post-weld heat-treat cracking.

Crack-Susceptibility Evaluation

Restraint Specimen Concept

A new crack-suscentibility specimen concept was developed and evaluated in the preceding contract effort[1]. This specimen was a circular-patch type utilizing a dissimilar alloy frame. The purpose of the dissimilar alloy frame was to provide a controlled augmented strain during the on-heating interval; this strain is the result of a difference in the coefficients of thermal expansion of the frame and disk alloys. The use of a frame material with a greater coefficient of thermal expansion serves as a means of superimposing a sustained biaxial state of stress on the specimen disk. (This specimen will be detailed in a following section of this report.) Thus, it is possible to select frame alloys which provide sufficient restraint stresses to produce cracking simulating varying restraint conditions experienced with production parts.

The circular-patch type specimen was selected for several reasons. It is representative of a complex weldment and subjected to a thermal and stress/strain history associated with an actual weld. These specimens are readily fabricated, somewhat symmetrical with respect to the resultant residual state of stress, and would, possibly, provide an opportunity for correlating results with those obtained by other investigators. The criteria established for this specimen included (1) the necessity that

cracking had to occur or originate at the inner test weld, (2) unaffected base metal cracking was not to be considered as part of the heat-treat cracking problem unless there was positive proof that cracking had originated at HAZ sites, and (3) correlation of cracking data would have to be based on definitive parameters such as the temperature at onset of cracking and crack location with all test variables, such as heating and cooling rates and temperature variations, fixed for each test. Results of the preceding program indicated that the previously noted criteria were attained and that the selected specimen configuration was (1) capable of reliably discriminating the nickel-base superalloy heat-treat cracking problem, (2) readily reproducible, (3) easily fabricated, and (4) economical to produce.

Acoustic Emission Analyses

An acoustic emission technique for monitoring cracking events during heat treatment was developed in the preceding contract effort to obtain data regarding the temperature dependence and nature of cracking. This technique was successful in overcoming one of the biggest shortcomings in previous heat-treat cracking studies; namely, the difficulty of obtaining quantitative information from the cracking process. The technique was found capable of distinguishing crack initiation and growth characteristics thus making it possible to establish acoustically defined parameters and quantifying crack-susceptibility results. The effects of aging contraction were readily revealed by acoustic emission analyses which permitted evaluation of the effects of contraction on subcritical cracking events. Correlations were made of emission characteristics and metallographic data of subcritical and gross cracks in Rene' 41. Acoustic signatures associated with subcritical cracking and slow and rapid crack growth were obtained for correlation purposes throughout the previous and subject programs. The selected acoustic emission test parameters were found to be responsive to variations in base-metals, pre-weld base-metal conditions, heat-treating procedures, and welding process parameter manipulations. Consequently, the acoustic emission technique was found to be an effective method of analysis and an invaluable laboratory tool for heat-treat cracking studies.

SECTION III

EXPERIMENTAL PROCEDURES

Materials

The chemical compositions of materials with which most of the work in this program was conducted are shown in Table I. These materials are left over from the previous program[1] thus permitting a correlation of results reported here with the prior work. Other heats of Rene' 41 and material from one heat of Waspaloy were also used for procedure verification purposes. The Rene' 41 Heat No's. 7470 and 6842 and Astroloy (similar to Udimet 700) are production-grades of the alloys. The REP powder bar stock was obtained as experimental material and converted to sheet at the Lockheed Georgia Research Laboratory for the previous contract effort. The general microstructures and other characteristics of these alloys may be found in Reference 1.

Welding and Heat Treating

All welding was performed with the gas tungsten-arc process using semiautomatic equipment and an automatic cold-wire feed system. A typical set-up for welding circular-patch crack-susceptibility specimens is shown in Fig. 1. Argon was used as the torch shielding gas and helium as the backing gas. Hastelloy W filler metal was used for both the inner and outer welds for all the materials involved in the program.

The heat treating procedures are the same as those developed previously[1] and involve the slow heating (15 to 17°F/min.) of test specimens to preselected temperatures. Differential heating effects were controlled with the use of fused silica foam insulation (see Fig. 2) to assure that the disk and frame temperatures were equalized prior to 900°F. This is a condition considered necessary to avoid thermal stresses due to temperature differences during on-heating through the aging temperature range. The use of thermal cycles involving nonuniform heating above 900°F would contribute to variations in the augmented thermal strains and thus would produce variations in results due to experimental rather than metallurgical effects.

TABLE I

CHEMICAL COMPOSITIONS OF MATERIALS USED IN THIS PROGRAM

Alloy	Heat Number	C	S	Mn	Si	Cr	Mo	Co	Ti	Al	B	Fe	Ni	Other	ASTM Grain Size
(1) Rene' 41[a] "Basic Material"	7470	.074	.008	.03	.08	18.75	9.68	10.92	3.36	1.56	.007	.51	bal.		7
(2) Rene' 41[b] (REP Powder Bar)	Ingot No. 8740-000001	.10	---	.05	.08	18.30	9.90	10.50	3.15	1.60	.009	1.95	bal.	O-.0036 N-.003	---
(3) Rene' 41[a]	6842	.075	.004	.05	.10	18.80	9.74	11.27	3.11	1.57	.005	1.17	bal.		7-8
(4) Astroloy[a]	6250	.056	.012	<.01	<.01	15.65	5.10	16.64	3.48	3.99	.023	.42	bal.	Zr <.02 Cu - .01 P - .003	10

Notes: (a) Analyzed and certified by Allvac Metals Company.

(b) Analyzed and certified by Whittaker Corporation, Nuclear Metals Division. The oxygen and nitrogen contents are those of the powder; the solids are the analyses of the original bar from which the powder was made.

FIGURE 1 - GAS TUNGSTEN-ARC WELDING SYSTEM USED FOR FABRICATION OF CRACK-SUSCEPTIBILITY TEST SPECIMENS

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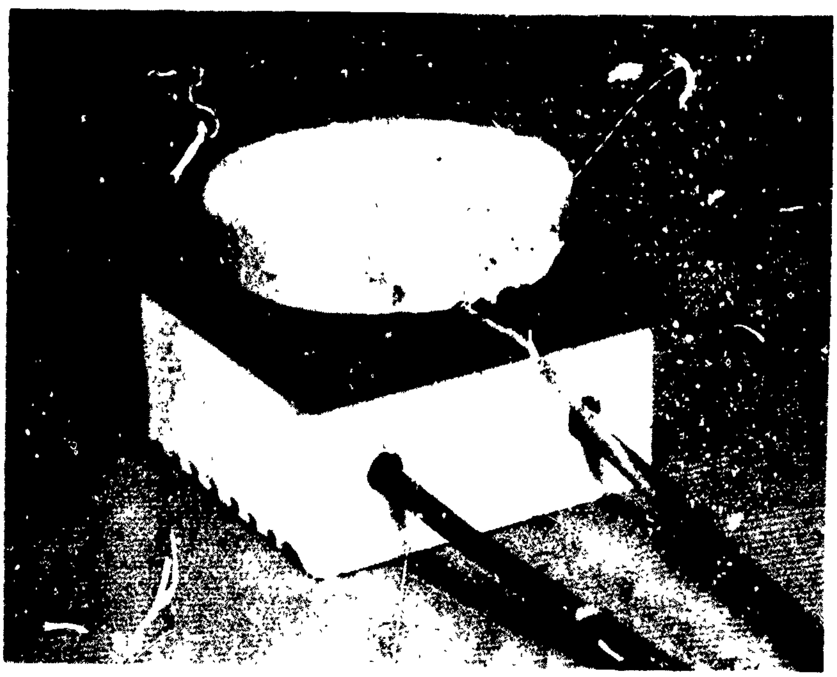


FIGURE 2 - HEAT-TREAT ENSEMBLE USED FOR ON-HEATING CRACK-SUSCEPTIBILITY TESTS. TONGS INSERTED IN BASE ARE USED FOR SPECIMEN TRANSPORT PURPOSES.

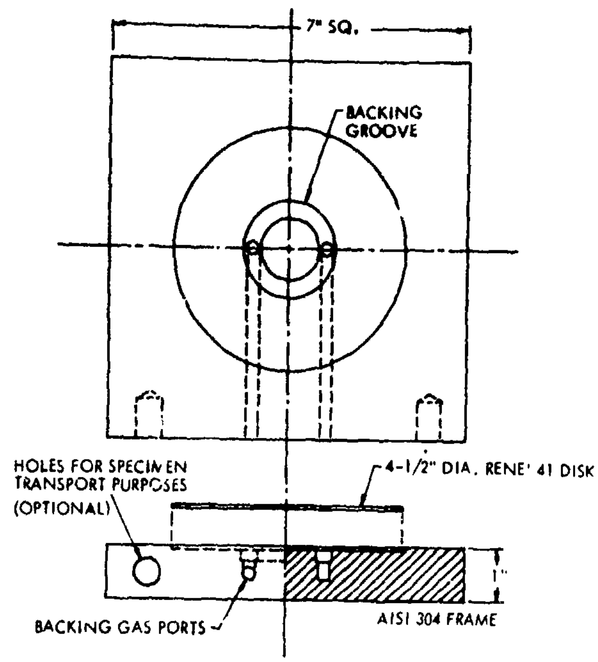
Restraint Test Specimens

The crack-susceptibility specimen configurations used in this program are shown in Fig. 3; on-heating tests were conducted with the specimen shown at top and isothermal tests with the specimen shown at bottom. Dissimilar alloy frames were used in all cases. Examples of fabricated assemblies are shown in Fig. 4 including an Astroloy disk/Inconel 600 frame "on-heating" specimen (left) and a Rene' 41 disk/ AISI 304 stainless frame "isothermal" specimen (right). The crack-susceptibility test criteria developed earlier(1) and briefly reviewed in the preceding section were used for all restraint specimen data obtained in this program. Examples of typical cracking obtained with these specimen configurations may be found in the preceding program report(1) and in subsequent sections of this report.

Acoustic Emission Monitoring During Heat Treatment

The equipment and techniques developed in the previous program(1) for monitoring cracking events during heat treatment were used for all crack-susceptibility tests conducted in this effort. The ability to monitor tests during heat treatment is attributed to the novel use of an "acoustic conductor" which permits coupling to a transducer outside of the furnace environment. The acoustic conductor is a small diameter wire, percussion welded to the specimen as shown in Fig. 5, which is extended through the parting interface of the furnace doors and coupled to a transducer outside the furnace. Consequently, this technique circumvents the transducer temperature limitation problem and permits the use of an analysis method which would, otherwise, not be practical.

The acoustic emission system assembled and used for this purpose is shown schematically in Fig. 6. The system used previously was replaced with a customized solid-state system featuring improved noise suppression characteristics, increased sensitivity, and greater latitude in signal processing. Acoustic signature characteristics of interest in this program included those associated with aging contraction occurring during on-heating through the aging temperature and micro and gross cracking events. The



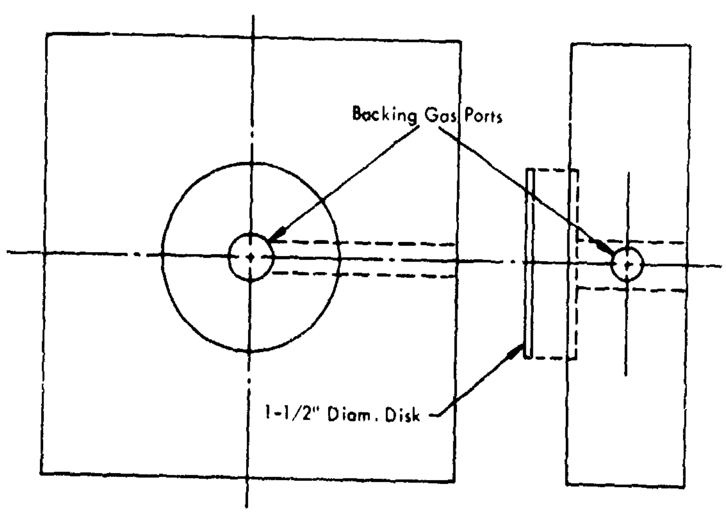


FIGURE 3 - ON-HEATING (TOP) AND ISOTHERMAL (BOTTOM) CRACK-SUSCEPTIBILITY SPECIMENS.

FIGURE 4 - TYPICAL FABRICATED ON-HEATING SPECIMEN (LEFT) AND ISOTHERMAL SPECIMEN (RIGHT)

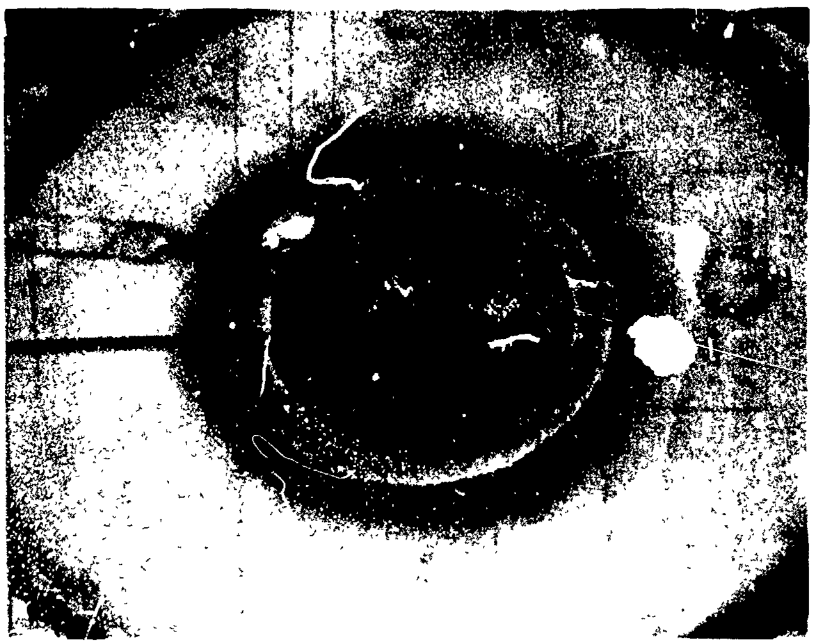


FIGURE 5 - TYPICAL INSTALLATION OF ACOUSTIC CONDUCTOR (LOWER WIRE) AND THERMOCOUPLE (UPPER WIRE) TO ON-HEATING SPECIMENS.

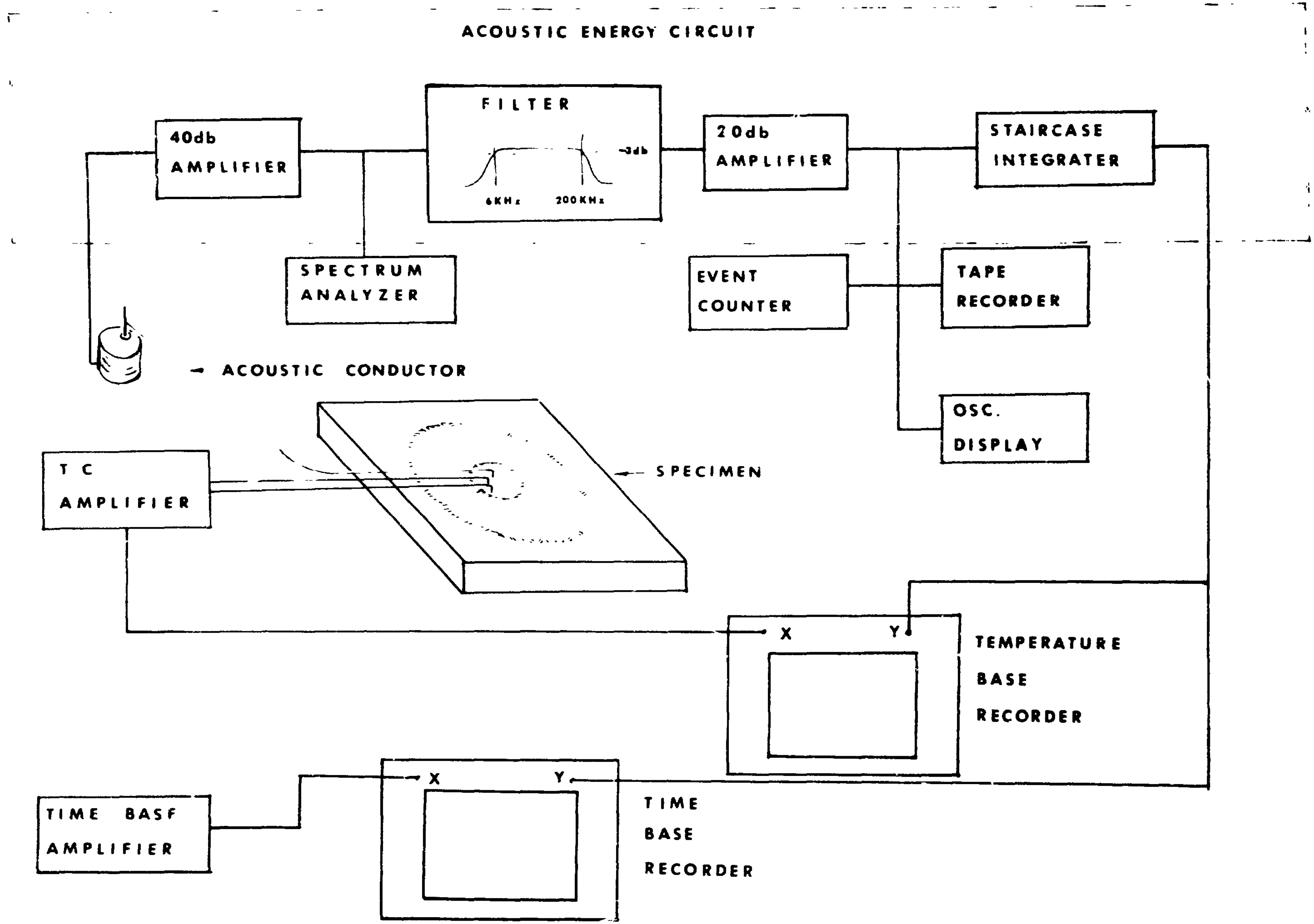


FIGURE 6 – DIAGRAMMATIC REPRESENTATION OF ACOUSTIC EMISSION SYSTEM USED FOR MONITORING HEAT-TREAT CRACKING

primary quantitative parameters used in this effort were the temperature at onset of gross cracking, noted as the "gross cracking temperature" from on-heating tests and the "time" at which cracking was experienced in isothermal tests. "Emission count" data and the "time interval of gross cracking," reported previously for these tests were found to be qualitative and are not reported; these parameters are directly reflected by the plots of acoustic energy as a function of temperature and time. Acoustic characteristics indicating the nature of crack growth are considered significant with respect to the crack resistance of the microstructure and are discussed for the tests conducted in this program.

SECTION IV
RESULTS AND DISCUSSION

Pre-Weld Solutioning Treatments

In the previous program[1], a number of pre-weld base metal heat treatments was studied as a means of "desensitizing" the HAZ microstructure. Of these heat treatments, a 1975°F-25 minute (water-quench) was found to be the most effective in avoiding on-heating cracking in Rene' 41. This solutioning treatment was also found to be effective for avoiding cracking in a low-carbon heat of Rene' 41 which was more crack sensitive (in the mill annealed condition) than the basic heat of Rene' 41 used for these studies. The use of this treatment was also found effective in raising the gross cracking temperature of Astroloy by 100 to 125°F.

The primary emphasis of this work was to explore, in more detail, the possibility of avoiding heat-treat cracking through use of optimized pre-weld solutioning treatments. This work was mainly concerned with the effects of varying solutioning times and temperatures within the 1925 to 2250°F temperature range and the effects of post-solutioning quench rate. The noted temperature range permits dissolution of gamma prime and the lower temperature carbides ($M_{23}C_6$); the higher temperature carbides (M_6C) are dissolutioned at temperatures at the upper end of this range (2150 to 2250°F). One of the objectives in these studies was to attempt to separate the influence of gamma prime and carbides on cracking behavior. This was to be accomplished by systematically varying the form of the carbide and noting its effect on cracking behavior during on-heating, that is, the interaction of carbides and gamma prime during and following aging contraction.

Two techniques used to study the effects of quench rate included water quenching of specimens heat-treated in Sen-Pak containers* and specimens heat-treated with "CR-THA" protective coating**. The Sen-Pak containers (stainless steel foil envelopes)

* Product of the Sentry Co., Foxboro, Mass.
**Product of the Markal Co., Chicago, Ill.

provided an effective air space barrier which served to lower the quench rate, whereas the "CR-THA" coating (a heavy suspension of vitreous material in a binder) which is liquid at solutioning temperatures functioned as an effective heat transfer medium and greatly increased the rate of quench. The difference in the quench efficiency of the two techniques is noted by the comparison of as-quenched hardness values; hardnesses of specimen in Sen-Pak containers quenched from 1975°F ranged from R_C 25 to 28, whereas hardnesses of specimens coated with "CR-THA" and quenched from 1975°F ranged from R_C 12 to 17. The significance of these quench rate variations is noted in later discussions.

Restraint Test Results

"On-heating" crack-susceptibility test results screening the various selected pre-weld heat treatments, along with the two noted post-solution quench rates, are summarized in Tables II and III. These data are for weldments in Rene' 41 (Heat No. 7470); the test results for the post-solutioned quenched material in Sen-Pak containers are noted in Table II and material with the "CR-THA" coating in Table III. These results clearly show that resistance to heat-treat cracking is influenced significantly by post-solution quench rate and to a lesser extent, but still significant, by the specific pre-weld solutioning treatment. Results shown in Table III indicate that the rapid quench achieved with use of the "CR-THA" coating is an effective method for avoiding gross or severe cracking during post-weld on-heating with all of the selected solution treatments. The results in Table II indicate that the lower quench rate technique, using Sen-Pak containers, is a less effective approach to avoiding gross cracking; this technique proved to be most effective with weldments in base metal given the pre-weld 1975°F solution treatments. Additionally, the incidence of microcracking in specimens which did not experience gross cracking was greater with weldments fabricated with the slower post-solution quenched base metal condition. Another trend evident with the results shown in Table II is that the "gross cracking temperature" increases with increasing solutioning temperature; from approximately 1360°F with the 1925°F treatments to as high as 1500°F with the 2250°F solutioning treatments. This trend is also consistent with the pre-weld hardness data for the respective solutioning temperatures. The lower hardnesses reflect a greater capacity for accommodation (relaxation) of restraint stresses during on-heating than do the higher pre-weld hardnesses.

TABLE II

ON-HEATING CRACK SUSCEPTIBILITY TEST RESULTS FOR RENE' 41 (HEAT NO. 7470) WELDMENTS FOR VARIOUS PRE-WELD SOLUTIONING TREATMENTS. MATERIAL HEAT-TREATED IN SEN-PAK[a] CONTAINERS

Pre-Weld Heat Treatment	Specimen No.	Temp., Onset of Gross Cracking (°F)	Hardness[b] Pre-Weld	Hardness[b] After Test
1925°F-1/2 hr., WQ	1(9)	1365		
"	2(10)	1360	34.0	39.0
"	3(11)	1355		
1925°F-4 hrs., WQ	4(48)	NGC[c]		41.0[e]
"	5(49)	1415	34.5	37.0
"	6(50)	1390		
1975°F-1 hr., WQ	7(12)	(d)	28.5	36.0[e]
"	8(29)	NGC[f]		
1975°F-4 hrs., WQ	9(13)	NGC[f]	25.0	33.5[e]
"	10(39)	NGC		
2050°F-1/2 hr., WQ	11(17)	NGC[f]		
"	12(26)	1435	27.0	3[illegible].0
"	13(27)	1425		
"	14(28)	1400		
2175°F-1/2 hr., WQ	15(18)	(d)		
"	16(19)	(d)	27.5	32.5
"	17(20)	NGC[f]		
2250°F-1/2 hr., WQ	18(51)	1470		
"	19(52)	1500	R_A 53.0	40.0
"	20(53)	1450		
"	21(54)	1425		

Notes: (a) Product of the Sentry Co., Foxboro, Mass.

(b) Hardness values in unaffected base metals; R_C unless otherwise noted.

(c) NGC - No Gross Cracking.

(d) Indicated specimens contained gross cracks unlike those usually obtained. The acoustic signatures for these cracks were typical of microcracking by predominantly slow (stable) crack growth.

(e) Includes 2-hr. dwell at 1400-1425°F.

(f) These specimens contained fine surface HAZ microcracks.

TABLE III

ON-HEATING CRACK SUSCEPTIBILITY TEST RESULTS FOR RENE' 41 (HEAT NO. 7470) WELDMENTS FOR VARIOUS PRE-WELD SOLUTIONING TREATMENTS. MATERIAL HEAT-TREATED WITH "CR-THA" COATING[a]

Pre-Weld Heat Treatment	Specimen No.	Temp., Onset of Gross Cracking (°F)	Hardness[b] Pre-Weld	Hardness[b] After Test
1925°F-1/2 hr., WQ	1(68)	NGC[c]	R_A - 60.5	40.0[d]
"	2(71)	NGC		
1925°F-4 hr., WQ	3(75)	NGC	R_A - 59.5	41.0[d]
"	4(76)	NGC		
1975°F-1/2 hr., WQ	5[e]	NGC	R_A - 57.5	41.0[d]
1975°F-4 hr., WQ	6(77)	NGC	R_A - 55.0	36.5[d]
"	7(78)	NGC		
2100°F-1/2 hr., WQ	8(69)	NGC[f]	R_A - 54.5	37.0[d]
"	9(79)	NGC		

Notes: (a) Product of the Markal Co., Chicago, Ill.

(b) Hardness values in unaffected base metals; R_C unless otherwise noted.

(c) NGC - No Gross Cracking.

(d) Includes 2-hr. dwell at 1400-1425°F.

(e) Summary of results of data from previous program reported in Technical Report AFML-TR-70-224 (October 1970).

(f) This specimen contained 3 microcracks less than 1/64" in length along the OD edge of inner weld.

These data are significant in that a strong dependence on γ' kinetics is suggested by the 1975°F solution temperature results. Previous work[1] also showed that a 1975°F-25 minute (air cool) treatment was also effective in avoiding gross cracking; however, some fine microcracks were encountered with this treatment. The data of Table II showing a tendency towards heat-treat cracking with pre-weld solutioning treatments above 1975°F could suggest a possible influence of the M_6C carbides and/or just simply grain growth. In any event, 1975°F is the most promising pre-weld solution temperature for avoiding heat-treat cracking.

Effects of Variations in Solutioning

The pre-weld hardness variations noted in Table II are undoubtedly direct evidence of the uniformity or completeness of solutioning. Hardness values decrease with increasing solutioning temperature; the hardness of the 4-hour 1975°F solutioned material is lower than the 1/2-hour treatment due simply to exposure time differences. The high pre-weld hardness values of the 1925°F treatments can be attributed to the fact that this temperature is just above the gamma prime dissolutioning temperature. The "after test" hardness values are indirect indications of the extent of hardening experienced during on-heating. The lower after-test hardnesses are generally associated with increasing gross cracking temperatures or with specimens which did not experience gross cracking. By comparison, the pre-weld hardnesses of the more rapidly quenched base metal specimens noted in Table III are substantially lower than those shown in Table II for all of the respective solutioning treatments. Again, as with the results noted in Table II, both pre-weld and after-test hardnesses decreased with increasing solutioning temperature. It should be noted, that the after-test hardnesses listed in Table III are the resultant values after a 2-hour dwell at 1400-1425°F and cannot, in all cases, be directly related to the foreshortened exposed specimen hardness results shown in Table II.

Specific microstructural effects of the pre-weld solution treatments noted in Tables II and III are clearly shown in Fig. 7. The microstructure of the 1925°F-1/2 hour material (top-left) is quite similar to that of the original mill annealed microstructure with grain boundaries delineated by precipitate and carbide networks and carbide

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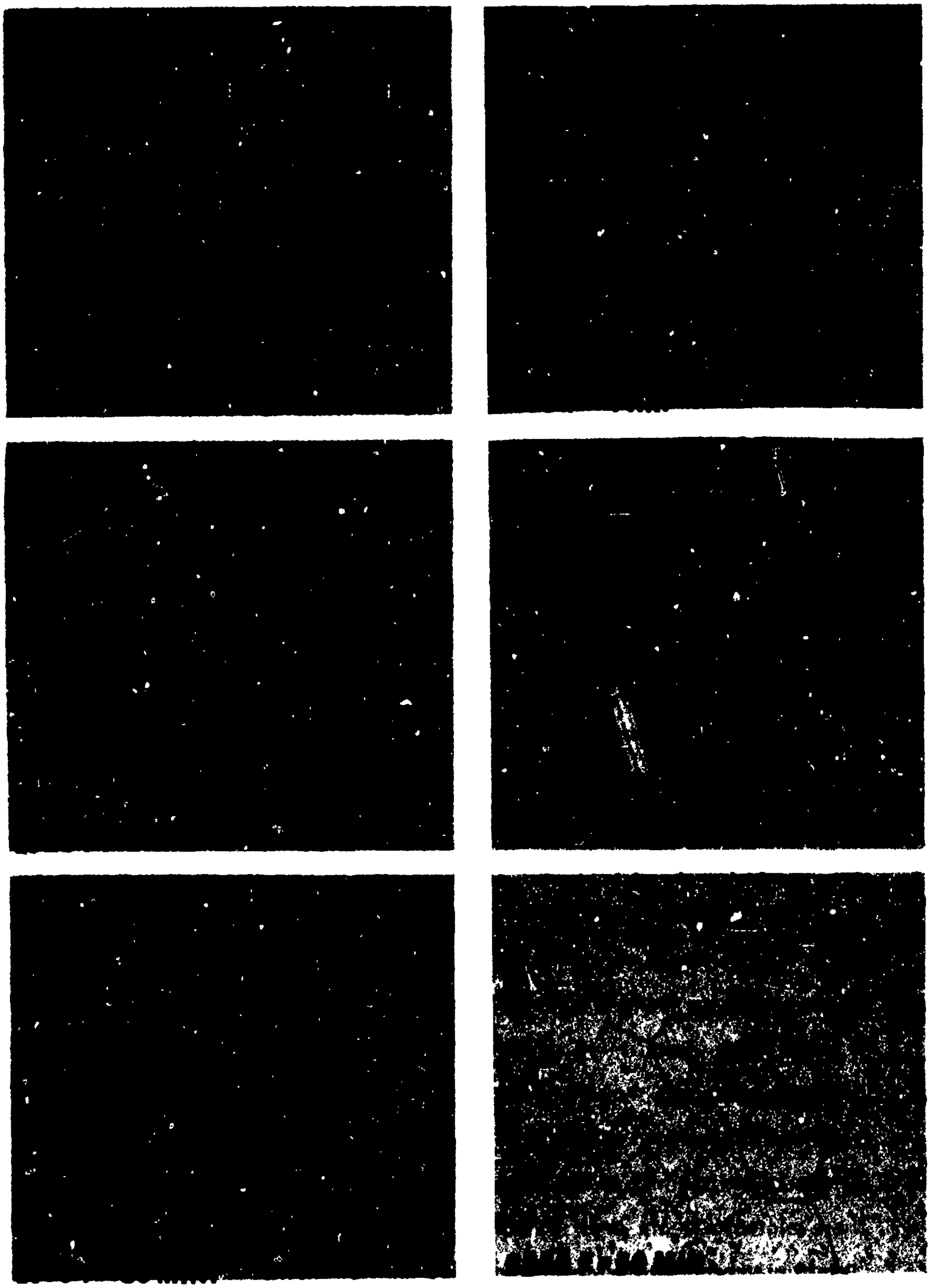
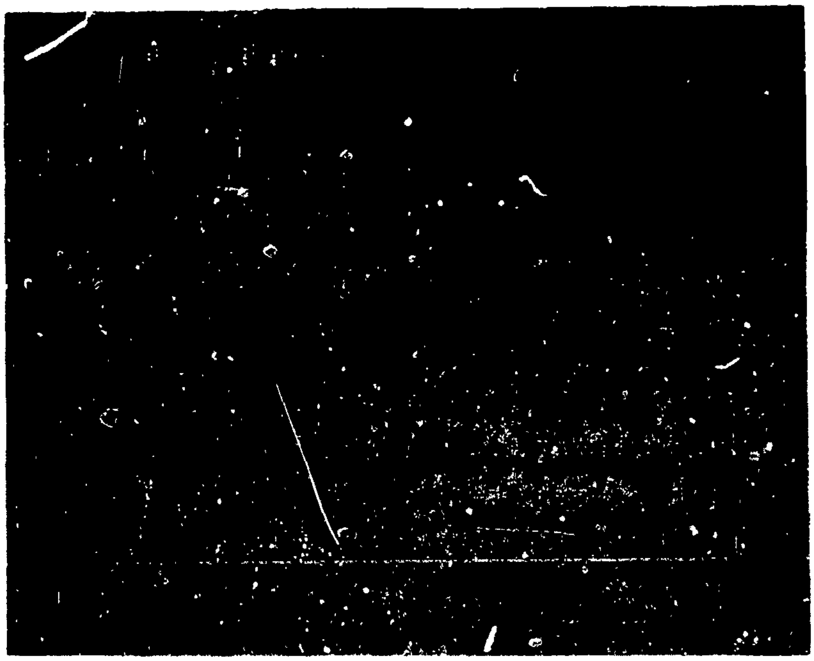


FIGURE 7 - RENE' 41 (HT. NO. 7470) BASE METAL MICROSTRUCTURES ASSOCIATED WITH THE VARIOUS PRE-WELD SOLUTION TREATMENTS USED IN THIS PROGRAM. ETCHANT: 60 HCl-30 LACTIC-7 HNO_3-3 H_2SO_4 -- X100.

stringers oriented with the rolling direction. The microstructure of the 1925°F-4 hour material (left-center) shows the apparent dissolutioning of the precipitate and carbide phases revealed by the partial delineation of grain boundaries; the carbide stringers are also affected but still apparent. The 1975°F-1/2 hour treatment (bottom-left) shows very little grain boundary structure indicating the more rapid dissolutioning of grain boundary precipitate and carbide networks; the residual carbide stringers are more apparent due to the lack of grain boundary delineated background. Dissolutioning of carbide(s) becomes more apparent with extended time at temperature (4 hours at 1975°F), as shown in the top-right photomicrograph and with increasing temperature (2100°F-1/2 hour) in the right-center photomicrograph; in both examples, the carbide stringers become progressively less evident. Finally, at the higher temperature (2250°F-1/2 hour) significant grain growth results, as shown in the lower-right photomicrograph. Contrary to the conclusions of some previous investigators, the results in Table II do not indicate a marked increase in the susceptibility to heat-treat cracking due to grain growth. The cracking data in Table II also suggests that the variation or scatter in gross cracking temperatures increases with increasing cracking temperatures.

The nature and extent of cracking associated with quench rate variations of the on-heating specimen results summarized in Tables II and III are worthy of note. Only one of the specimens in the Sen-Pak container quenched group did not exhibit cracking of any kind, whereas only one specimen of the more rapidly quenched "CR-THA" coated series was found to contain microcracks. Typical microcracks in the specimens of Table II are shown in Fig. 8 for specimens 8(29) and 17(20). This figure contains microstructural cross sections through crack indications revealed by fluorescent dye penetrant inspection. In both cases, cracking is intergranular and located at or in the immediate vicinity of the weld-heat affected zone juncture (toe region) of surface sites. These microcracks are quite typical to those observed earlier in this program with "interrupted on-heating" specimens used to correlate cracking characteristics with acoustic emission features.



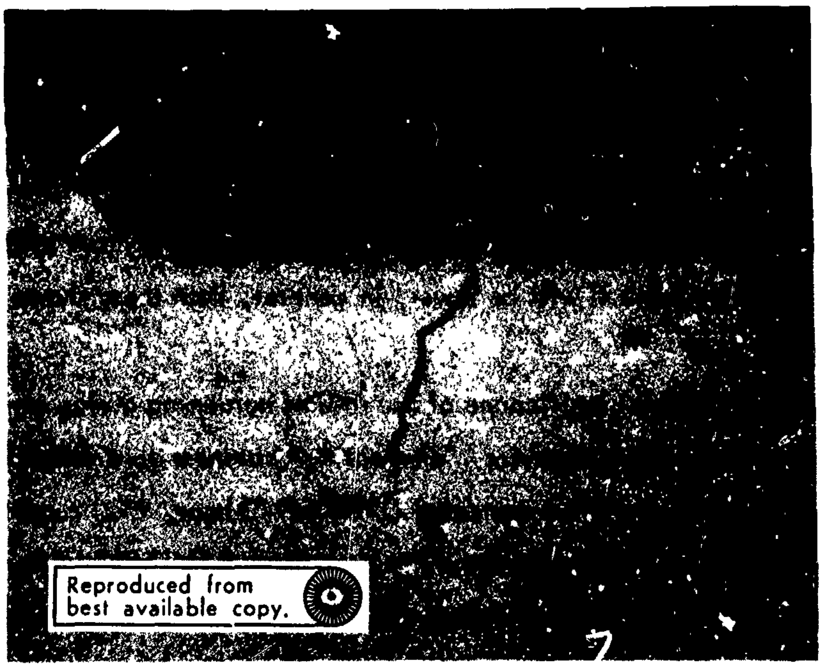


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FIGURE 8 - HEAT-TREAT MICROCRACKING IN PRE-WELD SOLUTION ANNEALED RENE' 41 (HT. NO. 7470) WELDMENTS. TOP: SPECIMEN NO. 8(29) -- X75, BOTTOM: SPECIMEN NO. 17(20) -- X415.

On-Heating Hardening Variations

The pre-weld and after test hardness data of Tables II and III indicate a marked difference in the rates of on-heating hardening and terminal hardness as a function of post-solution quenching and solution temperature. The more rapidly quenched conditions noted in Table III experience a greater change in hardness during on-heating (compare results of 1975°F solution series). This effect can be explained by considering the stage (extent) of precipitation associated with the pre-weld condition. The most effective quench should produce the most metastable structure in which pre-precipitation events are suppressed during quenching and would occur during the on-heating interval. The lesser effective quenches (revealed by higher pre-weld hardnesses) produce structures in which pre-precipitation events and, in many cases, initial stages of precipitation occur during post-solution processing. During on-heating, the most effectively quenched structures will experience the most rapid rate of hardening because the critical precipitate size is established sequentially and "clustering" conditions are ideal. The lesser effectively quenched structure will experience a lower rate of hardening due to the existence of clusters and/or precipitate particle sizes established for various temperatures during post-solution cooling. Hardening during on-heating will depend on the rate at which critical particle size conditions are developed for the respective temperatures; this is a time-temperature process and, consequently, the rate of hardening and the maximum (after test) hardnesses as shown in Table II will be lower, in general, than those shown in Table III.

Pursuant to the possible significance of the rate of hardening during on-heating and its relation with cracking behavior, "on-heating" hardness data were obtained for the mill annealed and solution annealed (1975°F-1/2 hour, WQ) conditions. These data were obtained with material from the basic heat (No. 7470) of Rene' 41 using heat-treat procedures identical to those used for the on-heating crack susceptibility tests. Specimen blanks were instrumented with thermocouples, placed on an AISI 304 stainless steel frame and covered with fused silica insulation in the same manner as with standard heat-treat tests. These specimens were heated (15-17°F per minute) to various temperatures ranging from 1000 to 1600°F. After arriving at the predetermined temperatures, the specimens were removed from the furnace, water

quenched, and then hardness tested. The hardness measurements thus obtained are plotted as the two curves shown in Fig. 9. These curves are plots of hardness variations during on-heating as a function of temperature; each point is the average of a minimum of four hardness measurements. The mill annealed condition curve plots above the solution annealed condition curve for obvious reasons. The higher initial hardness for the mill annealed condition is direct evidence of an advanced stage of precipitation and, consequently, gives rise to an earlier hardening response. As the solution annealed material results indicate, hardening per se does not commence until about 1150°F. The rate of hardening as noted by the slope of the curve is, however, greater than that exhibited by the mill annealed condition curve.

A point of note with regard to the curves of Fig. 9 is the indicated "temperature at onset of gross cracking" on the mill annealed plot and the corresponding hardness value on the solution annealed plot with respect to the on-heating temperature. Although the rate of hardening is greater for the solution annealed condition, the apparent critical hardness is not attained until about 1550-1600°F. Thus pre-weld solutioning treatments appear to be effective in avoiding heat-treat cracking through the simple expedient of "slack hardening" which permits plastic accommodation by the unaffected base metal of superimposed restraint, residual, thermal, and aging contraction stresses while the yield strength is still reasonably low. As indicated in Fig. 9, any shift downward of the on-heating hardness plot should denote a decreasing susceptibility to post-weld heat-treat cracking. This approach is consistent with the use of pre-weld overaging treatments wherein a flat on-heating hardness curve is obtained below the critical hardness; heat-treat cracking is avoided in the same manner as noted for the pre-weld solution treatments.

Effects of Pre-Weld Cold Work

In the previous program[1], it was found that 5 percent post-solution cold work reduced the effectiveness of the 1975°F-1/2 hour (WQ) solution anneal and resulted in on-heating gross cracking. Consequently, it was considered necessary to determine if small amounts of cold work, more representative of sheet forming operations, would

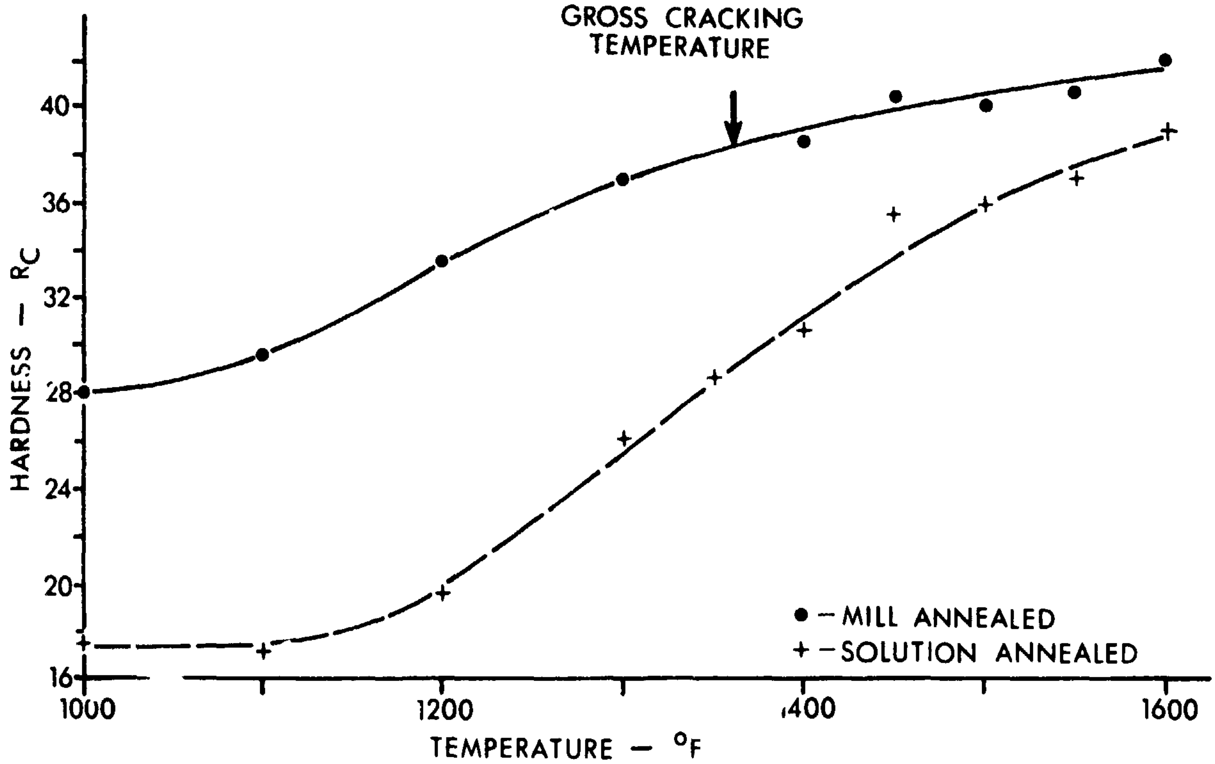

FIGURE 9 - "ON-HEATING" HARDNESS CURVES FOR MILL ANNEALED AND SOLUTION ANNEALED (1975°F-1/2 hr., WQ) RENE' 41 (HT. NO. 7470) BASE METAL.

also reduce the effectiveness of this pre-weld solution treatment for avoiding heat-treat cracking. For this purpose, a simple 2 percent cold reduction was selected for evaluation. Conventional crack-susceptibility specimens were fabricated using Rene' 41 (Heat No. 7470) disks punched from 5" x 10" sheets which had been solutioned (1975°F-1/2 hr, WQ) and cold rolled approximately 2 percent.

Results from on-heating tests conducted with these specimens are summarized in Table IV. The data indicate that the small amount of cold work introduced is not sufficient to cause either gross cracking as obtained previously with 5 percent cold work or microcracking during heat treatment. The significance of these results is that small amounts of cold work introduced during sheet forming operations should not increase the susceptibility of a pre-weld solution annealed component to post-weld heat-treat cracking. These data indicate that prior critical cold work is between 2 and 5 percent; the critical work being the amount of cold work which is sufficient to cause cracking during subsequent heat treatment. The effect of cold work is to promote or accelerate precipitation events earlier during the on-heating interval thus limiting the time for and/or capacity of the microstructure to plastically accommodate the imposed stresses.

Effects of Oxidation During Welding

One of the questions relative to heat-treat cracking of the superalloys is that of the possible influence of oxidation damage occurring during welding due to inadequate shielding. Taking advantage of the fact that crack initiation with the on-heating specimen used in this program occurs at the inner (root) surface of the specimen, three specimens were fabricated without the use of backing gas to permit oxidation of the root surface of the weld. On-heating test results with these specimens are contained in Table V. The gross cracking temperatures recorded for these specimens are varied, ranging from 1320 to 1360°F, which is a greater scatter than the ± 10 degrees experienced previously with the pre-weld mill annealed condition of the subject heat of Rene' 41. It should be noted, however, that the gross cracking temperature of specimen 1(22) is the only one that fell outside the range of gross

TABLE IV

"ON-HEATING" CRACK-SUSCEPTIBILITY TEST RESULTS SCREENING EFFECTS OF PRE-WELD SOLUTION ANNEAL PLUS 2% COLD WORK ON HEAT-TREAT CRACKING BEHAVIOR OF RENE' 41, HEAT NO. 7470

Specimen No.	Pre-Weld Base Metal Processing[a]	Temp., Onset of Gross Cracking (°F)	Hardness Data[b] Pre-Weld	After Test	Remarks
1(80)	1975°F-1/2 hr. (WQ) + CR	NGC[c]			CR-0.0647" to .0636"
2(81)	↓	NGC			CR-0.0645" to .0633"
3(82)	↓	NGC	R_A-59.0	41.5[d]	CR-0.0635" to .0620"
4(85)	↓	NGC			CR-0.0645" to .0633"
Reference Data[e]	1975°F-1/2 hr. (WQ)	NGC	17.0	41.0[f]	—
	1975°F-1/2 hr. (WQ) + 5% CR	1370, 1395	27.0	42.0	—

Notes:
(a) WQ - Water Quenched; CR - Cold Rolled.
(b) Hardness value in unaffected base metal; R_A unless otherwise noted.
(c) NGC - No Gross Cracking.
(d) Includes 2-hour dwell at 1400-1415°F.
(e) Data from Technical Report AFML-TR-70-224 (October 1970).
(f) Includes 1-hour dwell at 1400-1415°F.

cracking temperatures previously obtained for the mill annealed pre-weld condition. Based on the limited number of tests reported, the effects of oxidation damage arising from inadequate shielding are not entirely apparent. At worse, the prior damage may contribute to increased scatter in results tending towards a lowering of the gross cracking temperatures.

TABLE V

CRACK-SUSCEPTIBILITY RESULTS SHOWING EFFECTS OF OXIDATION DURING WELDING

Specimen No.	Material and Pre-Weld Condition	Temp., Onset of Gross Cracking (°F)	Remarks
1(22)	Rene' 41, Ht. No. 7470, (Mill annealed)	1320	Welded without backing gas
2(23)	↓	1350	↓
3(24)	↓	1360	↓
Reference[1]	Rene' 41, Ht. No. 7470 (Mill annealed)	1360	—

Effects of Furnace Atmosphere

The work described here was directed towards establishing the specific effects of heat-treat environment on post-weld cracking. The preceding program efforts[1] were inconclusive in this regard and tangible results with respect to reducing the incidence of heat-treat cracking through atmosphere control were not obtained. However, acoustic emission analyses indicated that variations in furnace atmosphere did influence the nature and extent of cracking. The specific effects of excluding oxygen from the furnace environment were not conclusively established and, consequently, it was considered necessary to conduct further tests to determine the influence of oxygen on heat-treat cracking. To accomplish this purpose, a new crack-susceptibility specimen configuration was developed which is a slight variation of

the on-heating restraint specimen used throughout this program. The new configuration is essentially an "integral retort" which permits precise control of heat-treat environment. The results described here are for tests conducted in air and in vacuum for the specific purpose of isolating the influence of oxygen.

Vacuum Test Specimen

The specimen configuration used for these tests is detailed in Fig. 10 and shown in Fig. 11 in an unassembled arrangement (top) and as a fabricated assembly (bottom). As seen in these figures, the specimen is circular and consists of three elements with the cover element serving the purpose of forming an enclosed chamber (integral retort). The disk and backing groove dimensions and welding procedures for this unit are the same as those of the standard square-shaped on-heating specimen used in this program. The two holes drilled from the top surface to the backing groove ports and located outside of the frame-to-disk weld provide access to the top of the chamber for evacuation purposes. The fabrication sequence for this specimen is essentially the same as that of the standard on-heating specimen. The outer disk-to-frame weld is made first and the inner (test) weld next. The cover is next attached with a single circumferential fillerless weld pass followed by the welding of inserted stainless steel dowels to the frame at one of the backing groove ports and the center positioning hole. The last operation involves the welding of the stainless steel evacuation tube assembly to the frame at the counterbored backing gas port. The evacuation tube was initially mechanically fastened to the frame, however, this approach was abandoned due to vacuum leak difficulties experienced during on-heating.

The cover element was intentionally made from plate stock to resist deflection during vacuum heat treatment. Recessing of the disk-frame top surface served several purposes: (1) contact of the inner and outer welds with the cover element was avoided, permitting flush contact of the cover and frame along the outer shoulder of the frame, (2) access holes to the backing groove ports could be located in the frame outside of the disk-to-frame weld, and (3) the recess served as a chamber, or cavity which permitted more effective evacuation of the space above the disk.

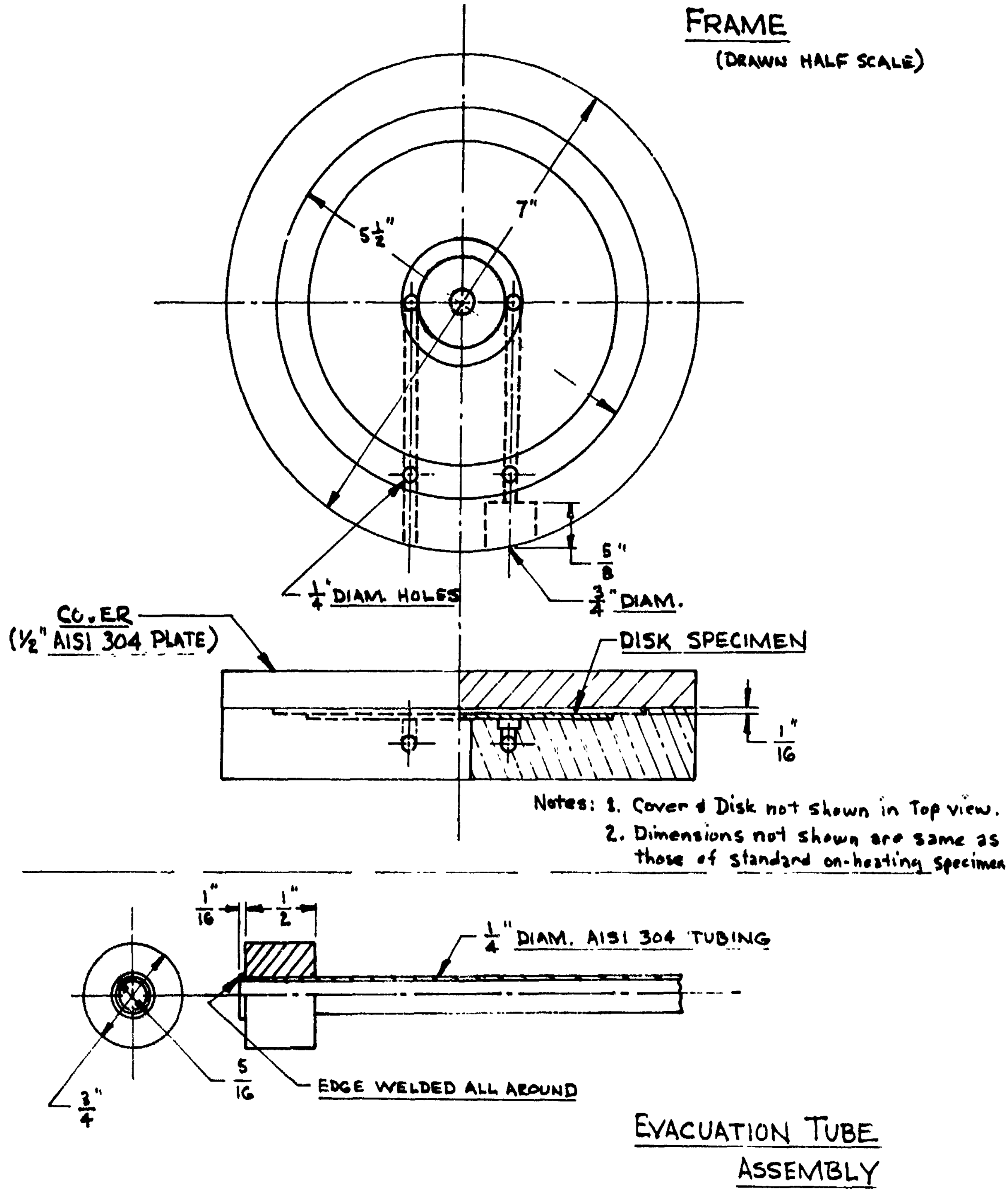

FIGURE 10 - ON-HEATING CRACK-SUSCEPTIBILITY SPECIMEN FOR HEAT-TREAT ATMOSPHERE STUDIES.

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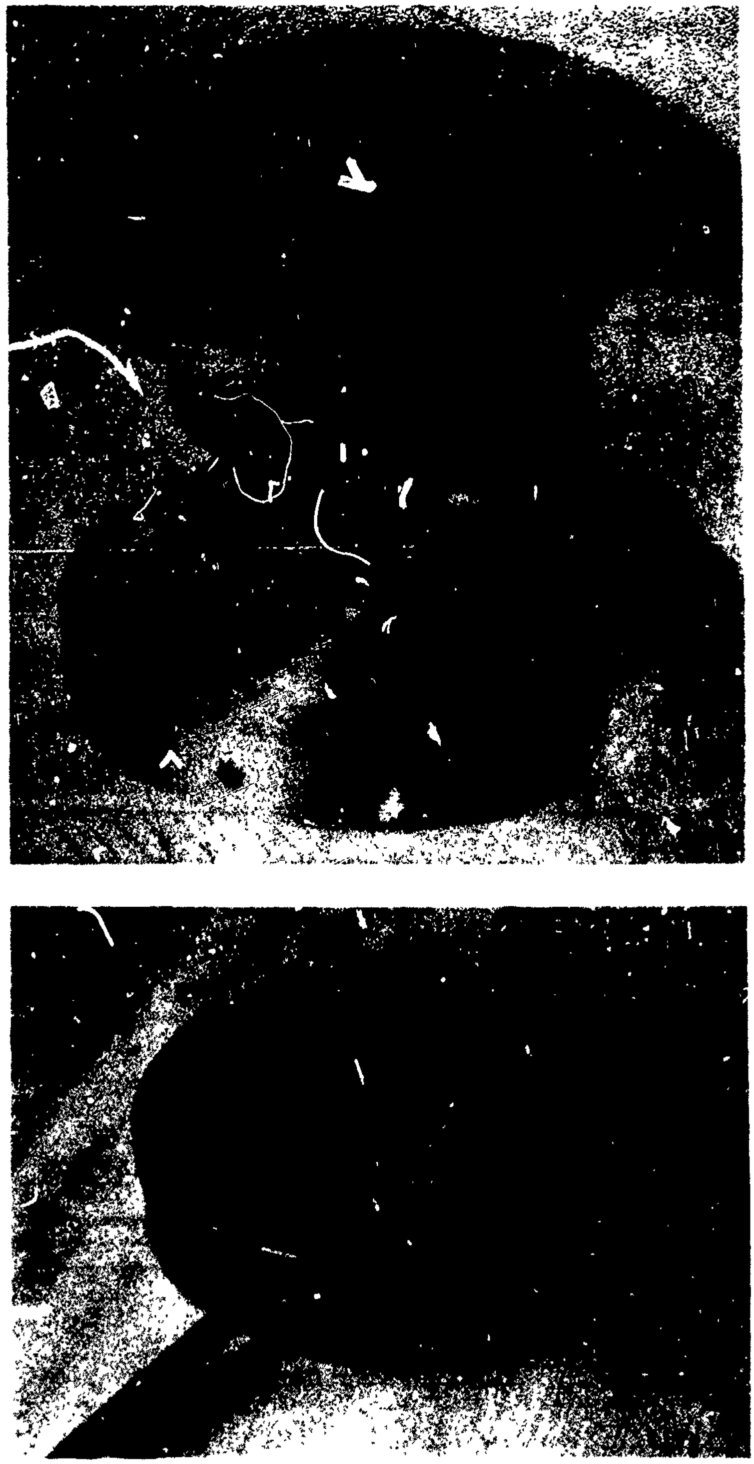

FIGURE 11 - ON-HEATING CRACK-SUSCEPTIBILITY SPECIMENS UNASSEMBLED (TOP) AND COMPLETELY FABRICATED (BOTTOM).

Test Procedures

The vacuum system used for these tests contained a 4" diameter diffusion pump which assured attaining partial pressures comparable to or better (lower) than those routinely obtained with production vacuum heat treating facilities. Coupling of the test specimen to the vacuum system was accomplished by attachment to the evacuation tube outside of the furnace. Specimen temperatures were obtained with thermocouples percussion welded to the center-top surface of the cover. One test using a specimen fabricated with a cover containing a hole through which a thermocouple was passed and installed to the center top-surface of the disk was conducted for the purpose of correlating the disk and cover temperatures during on-heating. Temperature variations above 900°F were found to be consistently within 10 degrees indicating that the cover temperature, at least for the air tests, was quite close to that of the disk insert. This small variation in temperature is probably attributable to the insulating effect of the fused silica foam "cold top" and "base" used for minimizing thermal shock and to achieve a slow heating rate during heat treatment (this heat treating procedure is detailed in Reference 1).

The procedures used for the vacuum specimen series included evacuation of the test specimen outside of the furnace until a low partial pressure was attained and then allowing for a pump-down period of at least two hours or overnight before conducting the heat-treat test. Instrumentation of the specimen was accomplished during the pump-down period. The acoustic conductor used for acoustic emission monitoring purposes was percussion welded to the surface of the cover-to-frame weld in the vicinity of the backing gas ports. The acoustic emission technique was the same as that used with the standard on-heating test series and, therefore, will not be elaborated upon here.

Crack-Susceptibility Test Results

Restraint test data obtained with the previously described circular-patch specimen are summarized in Table VI and VII. Data obtained for control purposes with air atmospheres are shown in Table VI and the vacuum environment test results are contained in

TABLE V.

"ON-HEATING" CRACK-SUSCEPTIBILITY CONTROL TEST RESULTS FOR RENE' 41 SPECIMENS HEATED IN AIR. DATA OBTAINED WITH CIRCULAR FRAME SPECIMEN SHOWN IN FIG. 10

Specimen No.	Base Metal	Pre-Weld Condition	Temp., Onset of Gross Cracking (°F)	Remarks
1(100)	Rene' 41 (Ht. No. 7470)	Mill Annealed	1375	Test conducted without cover (Figs. 10 & 11)
2(101)	↓	↓	1360	
Reference Data[(a)]	↓	↓	1350 - 1370	Reference data from standard square frame specimen used in this program
3(102)	Rene' 41 (Ht. No. 6842)	Mill Annealed	1335	
4(109)	↓	↓	1350	
5(110)	↓	↓	1340	
Reference Data	↓	↓	1335 - 1340	Reference data from standard square frame specimen used in this program

Notes: (a) Data from Technical Report AFML-TR-70-224 (October 1970) obtained with the standard square frame specimen.

TABLE VII

"ON-HEATING" CRACK SUSCEPTIBILITY TEST RESULTS FOR RENE' 41, HT. NO. 6842, SPECIMENS HEATED IN A VACUUM ENVIRONMENT. TEST SPECIMEN CONFIGURATION SHOWN IN FIGS. 10 & 11

Specimen No.	Pre-Weld Condition	Temp., Onset of Gross Cracking (°F)	Vacuum Pressures (mm Hg)[b] Initial	Vacuum Pressures (mm Hg)[b] Maximum	Remarks
1(104)	Mill Annealed	1360	1×10^{-4}	1×10^{-4} [c]	Inadequately cleaned frame[a].
2(105)		1350	3.5×10^{-5}	4×10^{-4}	Holes to upper chamber omitted.
3(106)		1355	1×10^{-5}	8×10^{-5}	
4(107)		1360	1.5×10^{-5}	8×10^{-5}	
5(108)		1340	1.8×10^{-5}	6×10^{-5}	
6(111)		1365	1.5×10^{-5}	5×10^{-5}	

Notes: (a) Specimen surface slightly discolored, however, crack faces appeared clean. This specimen was not ultrasonically degreased and residual machining oil probably remained in the backing groove parts which were rough surfaced and burred.

(b) The maximum pressure noted here is that observed in the temperature range above 900°F, during which subcritical and cracking events occur. Pressure increases below 900°F were attributed to surface out-gassing effects.

(c) Pressure surge to 4×10^{-4} mm Hg occurred during gross cracking when through-thickness cracking resulted causing evacuation of upper chamber.

Table VII. The air atmosphere tests were conducted with specimens of two heats of Rene' 41 (Heats 7470 and 6842) and the vacuum data with material from Rene' 41 Heat No. 6842. The data of Table VI provide a correlation of air atmosphere results obtained with the specimen configuration shown in Figs. 10 and 11, and the standard square frame specimen used throughout this program. As these data indicate, there is essentially no difference in the gross cracking temperatures for the two heats of material studied. This would suggest that air atmosphere data obtained with the standard frame specimen are valid for correlation purposes with the vacuum test data obtained with the circular frame integral-retort specimen.

A comparison of the gross cracking temperatures obtained with the vacuum environment tests summarized in Table VII with the air atmosphere results in Table VI clearly show that the use of a "good" vacuum for heat treating purposes will not significantly increase the resistance to post-weld heat-treat cracking. The noted 10 to 15 degree increase in the gross cracking temperature is of the same order of magnitude as the increases observed earlier by simply lowering the weld energy input. It is important to note here of the possibility of a temperature difference between the specimen disk and the top surface of the cover. With the vacuum tests, heating of the disk insert is accomplished by conduction through the base metal, whereas the air atmosphere specimens are additionally heated by the air in contact with the disk surfaces. Consequently, the disk insert temperature would probably lag the temperature of the cover element surface. If this thermal differential actually exists, it might be possible to conclude that a vacuum environment has no effect on gross cracking because the actual gross cracking temperature would be slightly lower than those noted in Table VII.

With regard to the quality of the vacuum used for these tests, attention should be drawn to the initial and maximum partial pressure data summarized in Table VII. Excluding the first two specimens for reasons noted in the table, initial pressures were quite low ranging from 1 to 1.8×10^{-5} mm Hg with maximum pressures of 5 to 8×10^{-5} mm Hg occurring in the 900°F to gross cracking temperature range. It is

significant to note that the gross cracking temperatures of the first two specimens shown in Table VII agree quite favorably with the gross cracking temperature obtained with the latter four specimens in this table. These data are in agreement despite the fact that the first specimen in this table was not adequately cleaned prior to heat treating and the second was fabricated using a frame in which the top-surface access holes were omitted.

Partial pressures measured during on-heating for each of the tests summarized in Table VII are plotted as a function of specimen temperature in Fig. 12. Of note in these plots is a characteristic increase in pressure which peaks-out in the 500 to 700°F temperature range. This effect is the result of surface out-gassing during heating and is clearly a function of the degree of cleanliness associated with the preparation of the specimens. For example, an effective initial low pressure could not be obtained with specimen 1(104) due to residual machining oil in the backing gas ports although the pressure did drop during on-heating above 900°F. Specimen 2(105), in which the top-surface access ports were omitted increased in pressure rapidly during on-heating and then declined during heating above 800°F; the surge in pressure at the gross cracking temperature is due to the sudden evacuation of the upper chamber resulting from the development of through-thickness cracks. The low partial pressure cycle exhibited by specimen 6(111) is the result of meticulous cleaning procedures used to verify the influence of out-gassing on partial pressure variations during on-heating. The primary difference in the preparation of this specimen as contrasted with the others was the belt grinding of the inside surface of the cover element to remove the mill surface material which was thought to be a predominant source of out-gassing.

Metallographic Analyses

The effectiveness of the vacuum environment is illustrated by the specimens shown in Fig. 13. The specimen on the left is an air atmosphere test and the one on the right is a vacuum test specimen. The disk placed on the two specimens illustrates the luster of the as-received material for comparison purposes. Although not evident

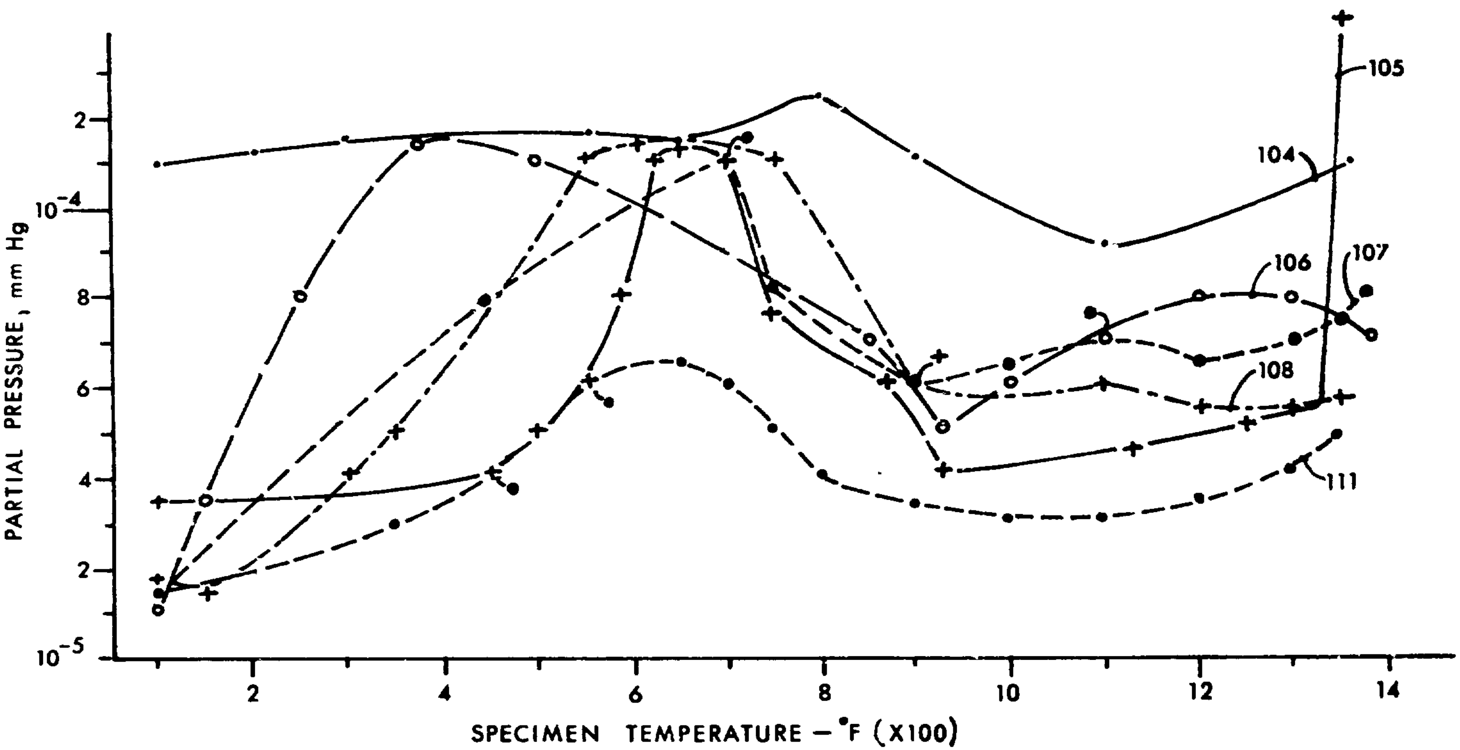

FIGURE 12 - ON-HEATING PARTIAL PRESSURE PROFILES OBTAINED FOR VACUUM TEST SERIES.

FIGURE 13 - COMPARISON OF SURFACE OXIDATION DUE TO AIR ATMOSPHERE (LEFT) AND VACUUM ENVIRONMENT (RIGHT). DISK IN CENTER IS AS-RECEIVED MATERIAL.

in this photograph, the surfaces of the disks from the vacuum tests appeared "cleaner" after heat treatment than they were prior to heat-treat cycling. The effects of oxygen were clearly revealed during the routine metallographic examination of test specimen fractures. These effects are shown in the scanning electron micrographs of Fig. 14, where selective intergranular oxidation of heat-affected zone surfaces is evidenced for varying oxygen content conditions. The micrograph on the left shows the oxidation attack of the heat-affected zone due to heat treating in air atmosphere. The center micrograph reveals less pronounced selective attack in the immediate vicinity of a gross crack obtained with an earlier test in which the air leak was obtained during on-heating. The micrograph on the right, taken in the immediate vicinity of a gross crack in a vacuum test specimen, does not show any signs of intergranular oxidation; the slight surface indications apparent in the micrograph are probably due to the opening of prior oxidation damage sites resulting from microstraining of the surface metal in the vicinity of the heat-treat crack network. The abrasion marks evident in all the examples of Fig. 14 are due to pre-weld sanding of the surfaces preparatory to welding.

A systematic scanning electron microscopy analyses comparing the fracture features of air atmosphere specimens with the features of vacuum specimens were conducted. The variations in fracture appearance between these groups of specimens were readily apparent. Examples of scanning electron micrographs comparing fracture features of air atmosphere and vacuum environment specimens are shown in Fig. 15. The micrographs to the left are the air atmosphere specimen features and the ones on the right are features of vacuum specimen fractures. The general oxidation of the air atmosphere test fractures is quite similar in appearance to fractures which have undergone general corrosion attack. In contrast, the vacuum specimen fractures are sharp featured and secondary cracking is more readily apparent.

A detailed replica microscopy study was also conducted to further establish microstructural feature differences associated with fractures in air and vacuum. Representative micrographs of fracture features associated with these furnace environments are shown in Figs. 16 (air) and 17 (vacuum). In both instances, fracture is

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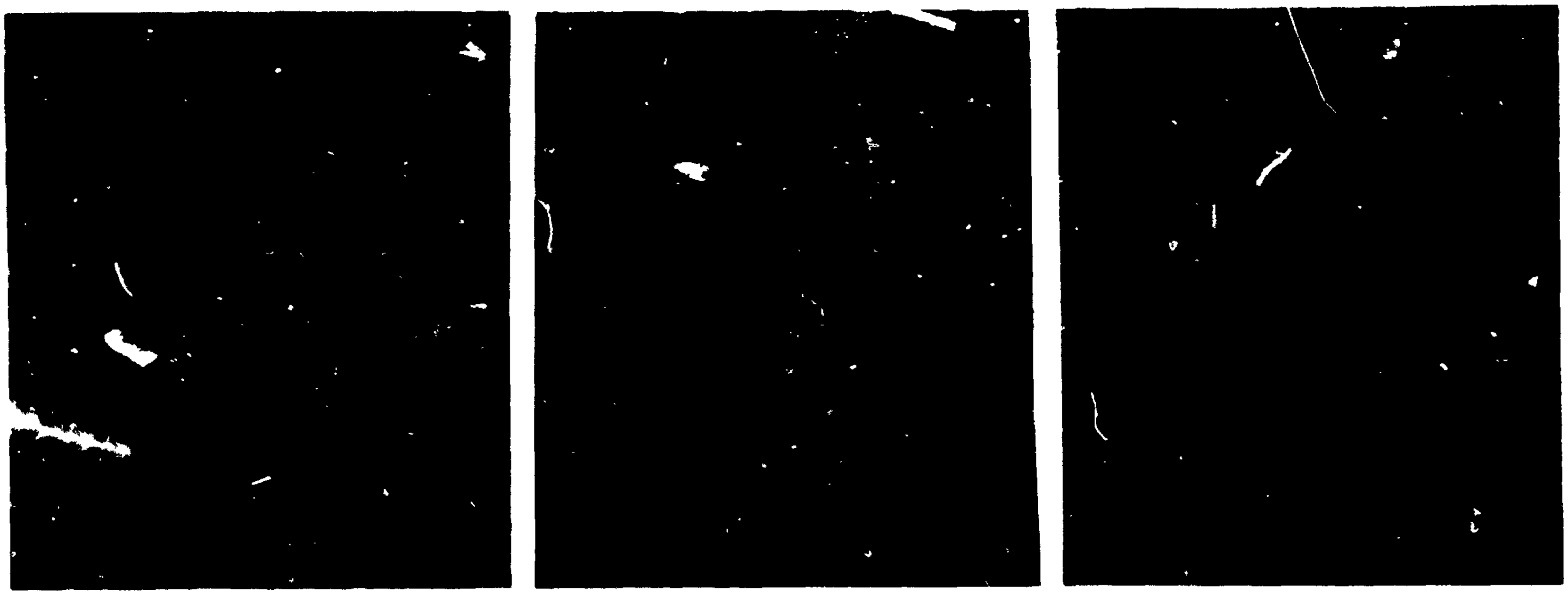

FIGURE 14 - SCANNING ELECTRON MICROGRAPHS SHOWING SELECTED INTERGRANULAR OXIDATION OF THE HAZ SURFACE DUE TO HEAT-TREATING IN AN AIR ATMOSPHERE (LEFT) AND A VACUUM ENVIRONMENT (RIGHT). CENTER EXAMPLE WAS TAKEN FROM A VACUUM SPECIMEN WHICH DEVELOPED A LEAK DURING ON-HEATING. X1000

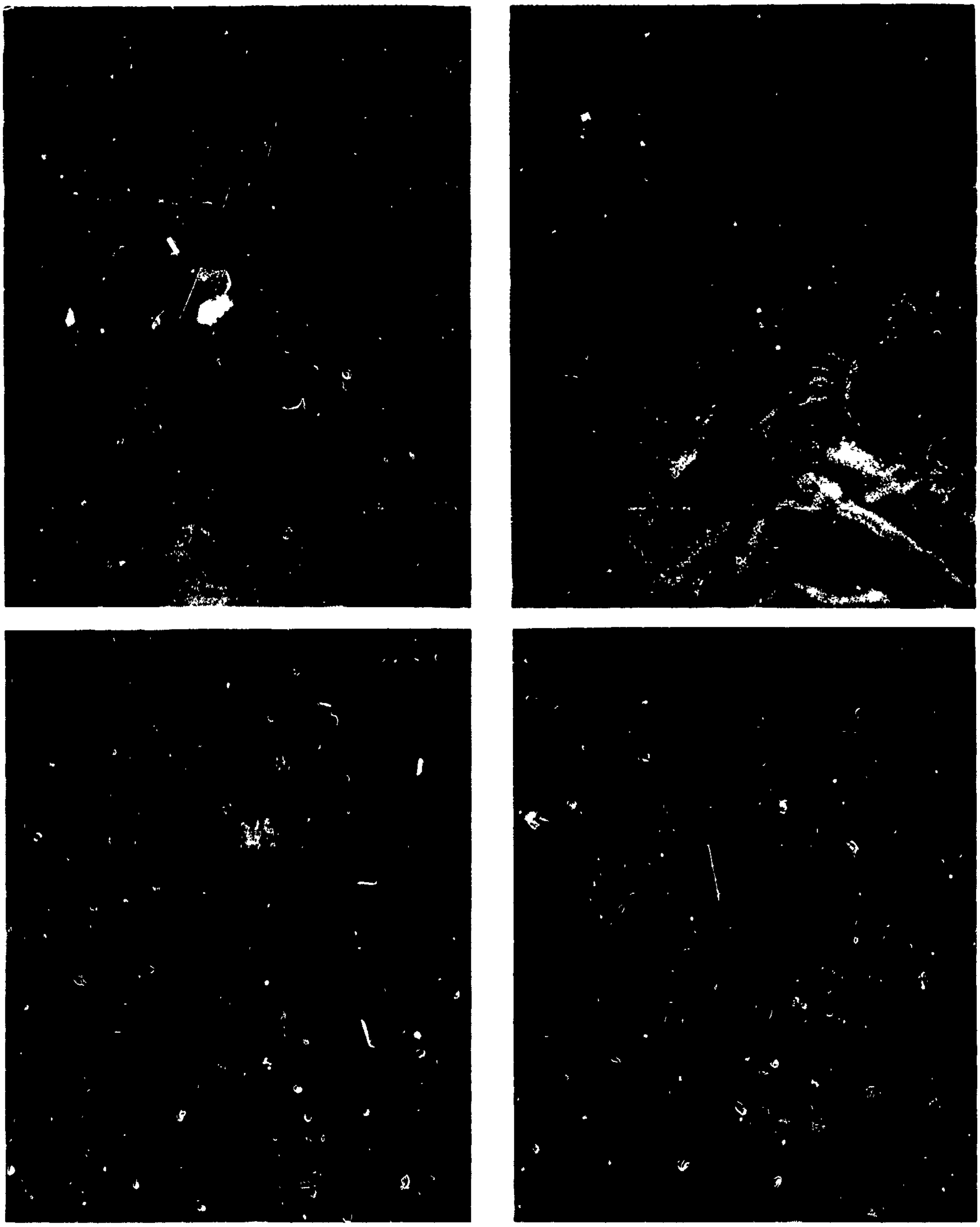

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FIGURE 15 - SCANNING ELECTRON MICROGRAPHS SHOWING DIFFERENCES IN FRACTURE FACE FEATURES DUE TO HEAT TREATING IN AN AIR ATMOSPHERE (LEFT) AND A VACUUM ENVIRONMENT (RIGHT). X1000

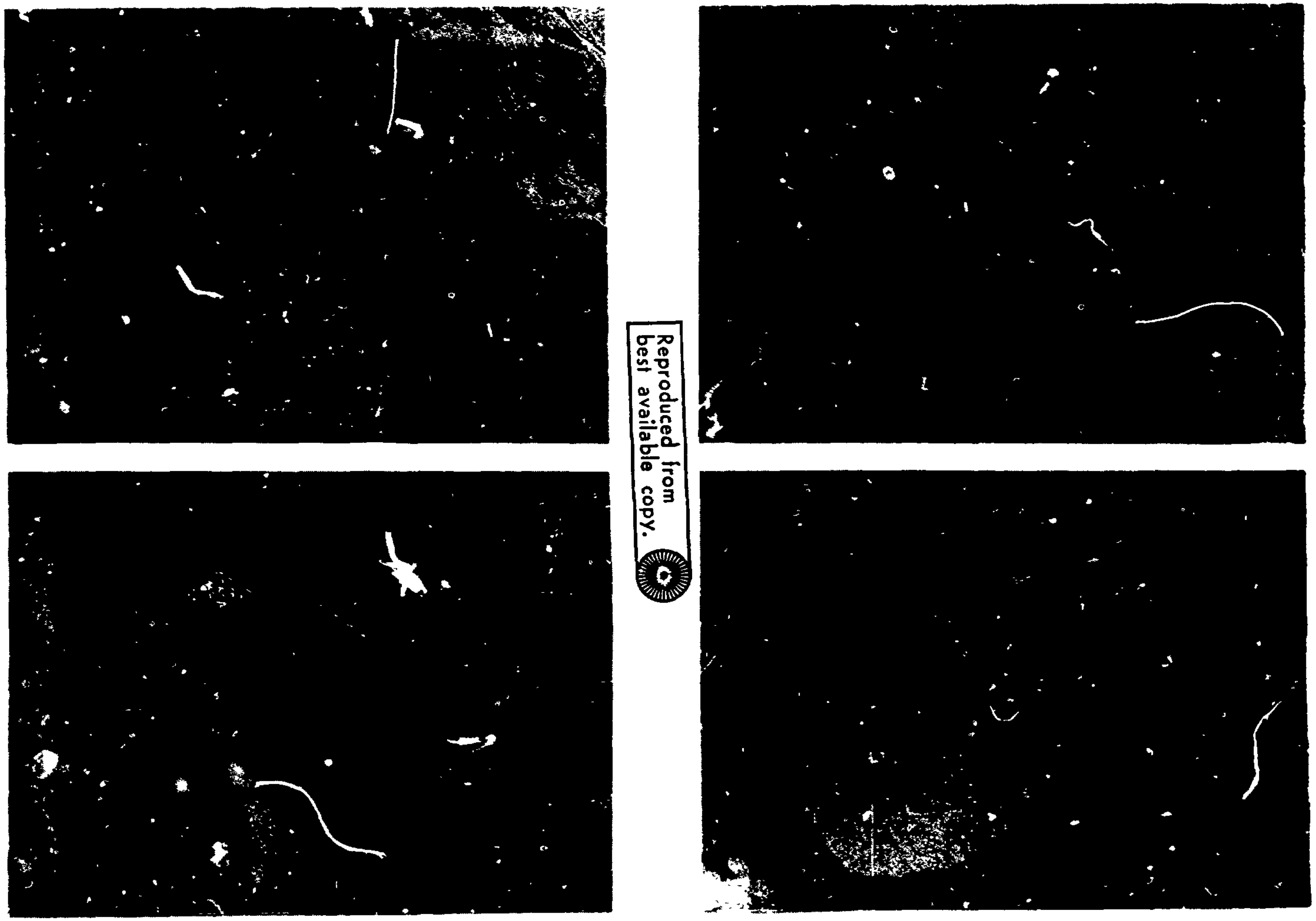

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FIGURE 16 - REPLICA ELECTRON FRACTOGRAPHS OF RENE' 41 POST-WELD HEAT-TREAT (AIR) CRACKING. TOP - X2000, BOTTOM - X10,000.

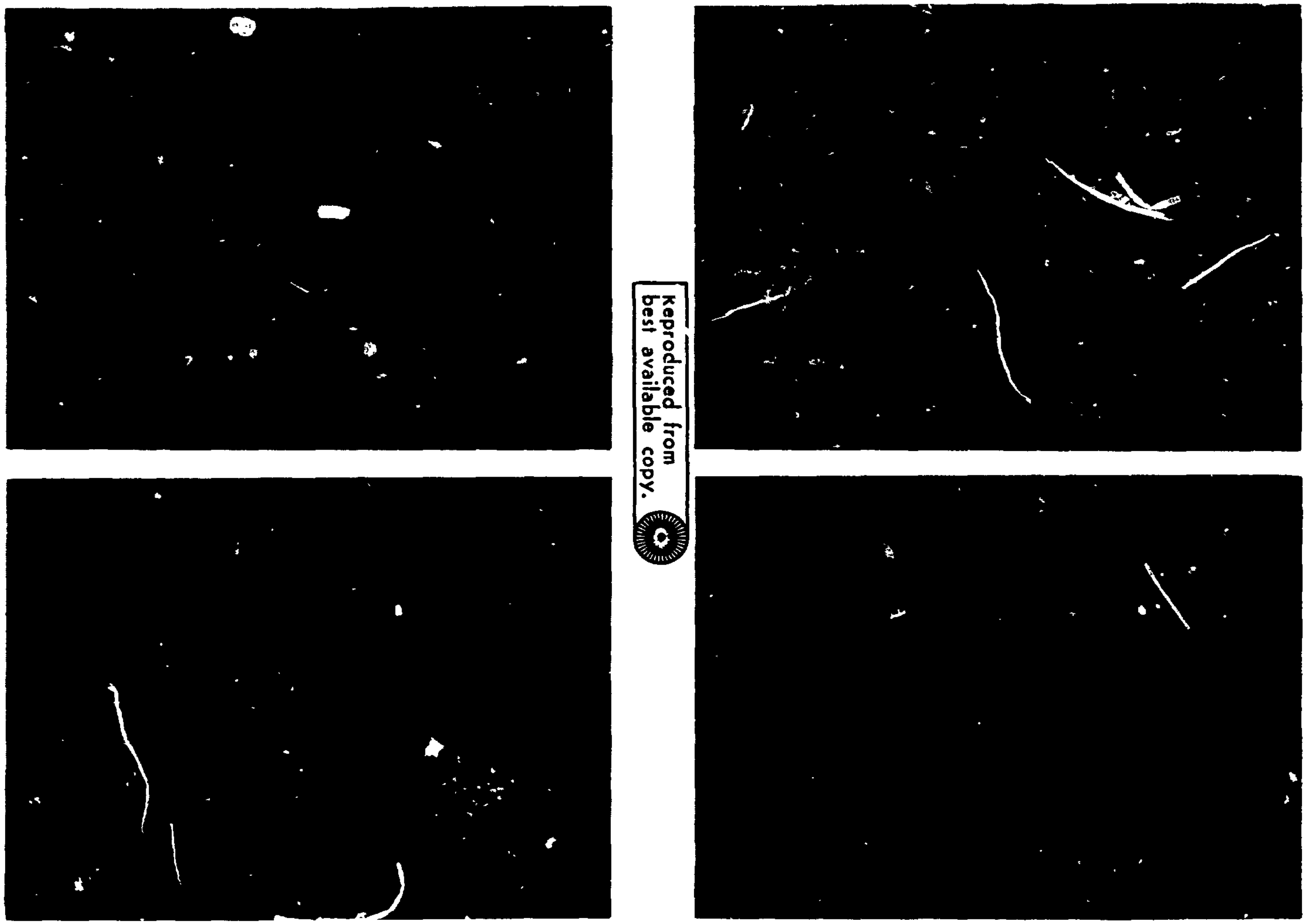

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FIGURE 17 - REPLICA ELECTRON FRACTOGRAPHS OF RENE' 41 POST-WELD HEAT-TREAT (VACUUM) CRACKING. TOP - X2000, BOTTOM - 10,000.

intergranular and compound particles are evident on the grain facets. The primary difference shown in these figures is the oxidation of micro-features such as compound particles and precipitate particle sites on the air atmosphere fractures. It is not apparent whether oxidation occurs as part of the fracturing process or just simply subsequent to fracture as the fresh surfaces are exposed to air. The restraint test results would suggest the latter since the difference in susceptibility to heat-treat cracking of air and vacuum specimens was not significant. Another fracture characteristic noted in this study was the common occurrence of "striated" separations with vacuum test specimen fractures. These striated features were also evident with air atmosphere fractures, but to a considerably lesser degree. Examples of the striated features associated with heat-treat cracking in vacuum are shown in Fig. 18.

The most likely explanation for the occurrence of these stria is that they are slip plane traces accommodating the fracture process. The possibility of these striations representing cyclic crack extension similar to fatigue crack growth was considered; however, the lack of any apparent continuity of the crack front among adjacent grains did not support this idea. The presence of these striations might be consistent with fractures exhibiting greater plasticity characteristics which is somewhat supported by the slower rate of crack extension revealed by acoustic emission analyses. It may also be assumed that step-like fracturing along grain boundaries indicates a higher resistance to crack extension than does a planar separation of grains. Consequently, it might be concluded that a vacuum environment is more desirous than the use of air atmospheres for heat-treating purposes; this is suggested by the restraint test results showing a slight increase in the gross cracking temperature using a vacuum environment.

Acoustic Emission Analyses

Significant differences were noted in the acoustic emission cracking characteristics as influenced by an air atmosphere and a vacuum environment. These differences were noted for both subcritical and gross cracking events during on-heating. The continuous emissions and occasional low and intermediate intensity bursts normally observed with

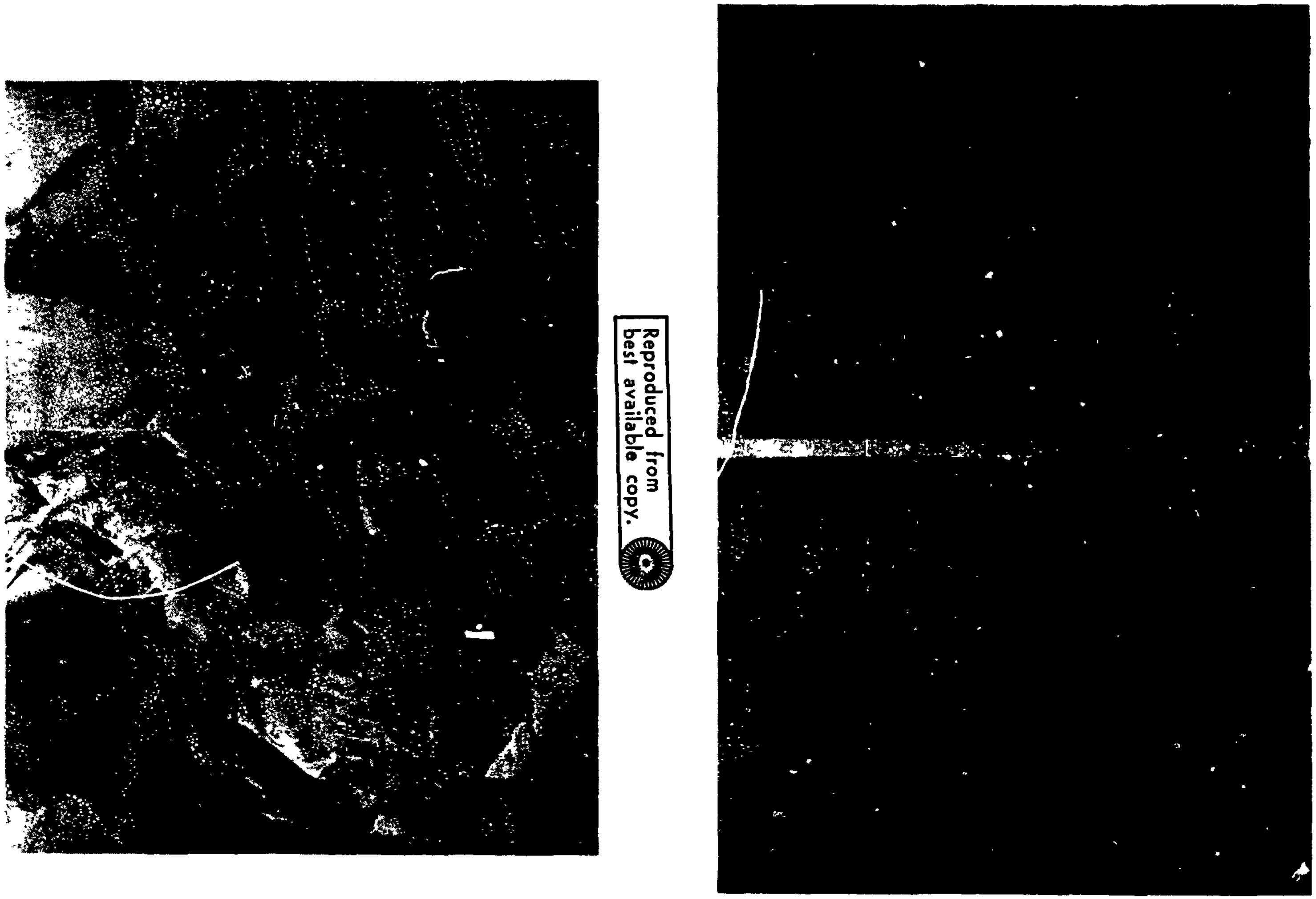

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FIGURE 18 - REPLICA ELECTRON FRACTOGRAPHS SHOWING STRIATED FEATURES IN RENE' 41 HEAT-TREAT (VACUUM) CRACKING.

air atmosphere tests were far less pronounced with the vacuum test series. This would suggest the possibility that this acoustic activity is associated with selective intergranular surface oxidation. Subcritical microcracking indications just prior to gross cracking were more evident with the vacuum test series than with air atmosphere tests. This latter effect may also be related to the lack of surface oxidation during vacuum heat treatment. It would be reasonable to assume that stress-aided intergranular oxidation contributes to the stress relaxation of surface layer material through propagation of a multiplicity of incipient surface cracks, especially at sites of prior oxidation penetration damage. Consequently, the residual surface layer stresses of the vacuum test specimens would be greater to higher on-heating temperatures causing microcracking to occur later in time as precipitation hardening progressively limits the plastic strain capacity of the bulk microstructure. The acoustic emission characteristics of gross cracking of air atmosphere and vacuum environment tests also varied significantly. Acoustic activity of vacuum specimens consisted of profusions of intermediate intensity continuous emissions with occasional acoustic bursts, whereas gross cracking of air atmosphere tests consisted of numerous acoustic bursts and much lesser amounts of continuous emissions. These variations in acoustic signatures would suggest that a vacuum environment increases the resistance to gross cracking by shifting the mode of crack extension from predominantly unstable crack growth in air to predominantly stable (slower) crack extension. This was somewhat verified by observations of specimen fractures in that some part-through cracking was found in the vacuum test specimens; these were the first observations of part-through gross cracking noted in this and the preceding program.

It was also observed that the gross cracking temperature intervals for the vacuum test specimen were always smaller than those for the air atmosphere tests which is another indication of increased resistance to gross cracking. In summary, the use of a vacuum environment appears to increase the resistance to heat-treat cracking, but does not appear to be a total approach to avoiding post-weld cracking. Vacuum heat treating does affect the nature and extent of cracking experienced during on-heating and appears to increase the gross cracking temperature. The use of a vacuum heat treating environment is, at least, as effective as lowering the weld energy input,

noted in previous work, but considerably less effective than the use of pre-weld solutioning for avoiding heat-treat cracking in Rene' 41.

REP Powder Sheet Evaluation

In the previous program[1], on-heating and isothermal crack-susceptibility test results for Rene' 41 sheet converted from REP (Rotating Electrode Process) powder bar stock indicated that the pre-weld solution annealed material exhibited a greater resistance to heat-treat cracking than did the pre-weld mill annealed basic heat (No. 7470). On-heating tests comparing the pre-weld solution annealed conditions of these materials did not show a difference in behavior since cracking was not observed with either material. A comparison of isothermal cracking behavior for the pre-weld solution annealed condition was not possible since isothermal tests with the basic heat in this condition were not conducted. Consequently, isothermal data were obtained in this program effort and are summarized in Table VIII.

REP material specimens 1 through 4 in Table VIII were tested previously and the results included for reference purposes. Specimen 5(4) was run to obtain a correlation with the previous results and was found to be consistent in that small tight "part-through" cracks were obtained during isothermal exposure. The pre-weld solution annealed basic material, specimens 6 through 8, survived isothermal exposure without cracking, indicating a greater resistance to heat-treat cracking than the REP powder sheet material. It should be noted that previous isothermal tests with pre-weld mill annealed basic heat specimens experienced gross cracking which indicates further the effectiveness of pre- veld solutioning for avoiding heat-treat cracking.

Attention should be drawn to the fact that cracks associated with the REP sheet specimens were "part-through" cracks indicating some measure of crack retardation capability. This inherent resistance to crack extension may be attributed to an in-plane micro-delamination tendency as shown in the photomicrograph of Fig. 19. The upper photomicrograph clearly illustrates delaminations presumably along

TABLE VIII

"ISOTHERMAL" CRACK-SUSCEPTIBILITY DATA COMPARING RENE' 41 SHEET CONVERTED FROM REP POWDER BAR AND CONVENTIONAL RENE' 41 WROUGHT SHEET (HEAT NO. 7470)[a]

Specimen No.[b]	Pre-Weld Base Metal Condition	Total Length of Crack(s) Observed[a]	Number of Cracks
REP Powder Sheet Specimens			
1(7)	1975°F - 25 min., WQ(R_C20)	1/4"	1
2(8)	"	5/32"	2
3(9)	"	1-3/8"	1
4(20)	"	21/32"	7
5(4)	"	1/8"	2
Wrought Sheet, Rene' 41 Heat No. 7470			
6(1)	1975°F - 1/2 hr., WQ (R_C19)	No cracks Detected.	
7(2)	"	"	
8(3)	"	"	

Notes: (a) All specimens exposed for 4 hours at 1400-1425°F.

(b) Data for specimens 1 through 4 obtained from Table XV of Reference 1.

(c) All cracking of REP powder sheet specimens were "part-through" cracks.

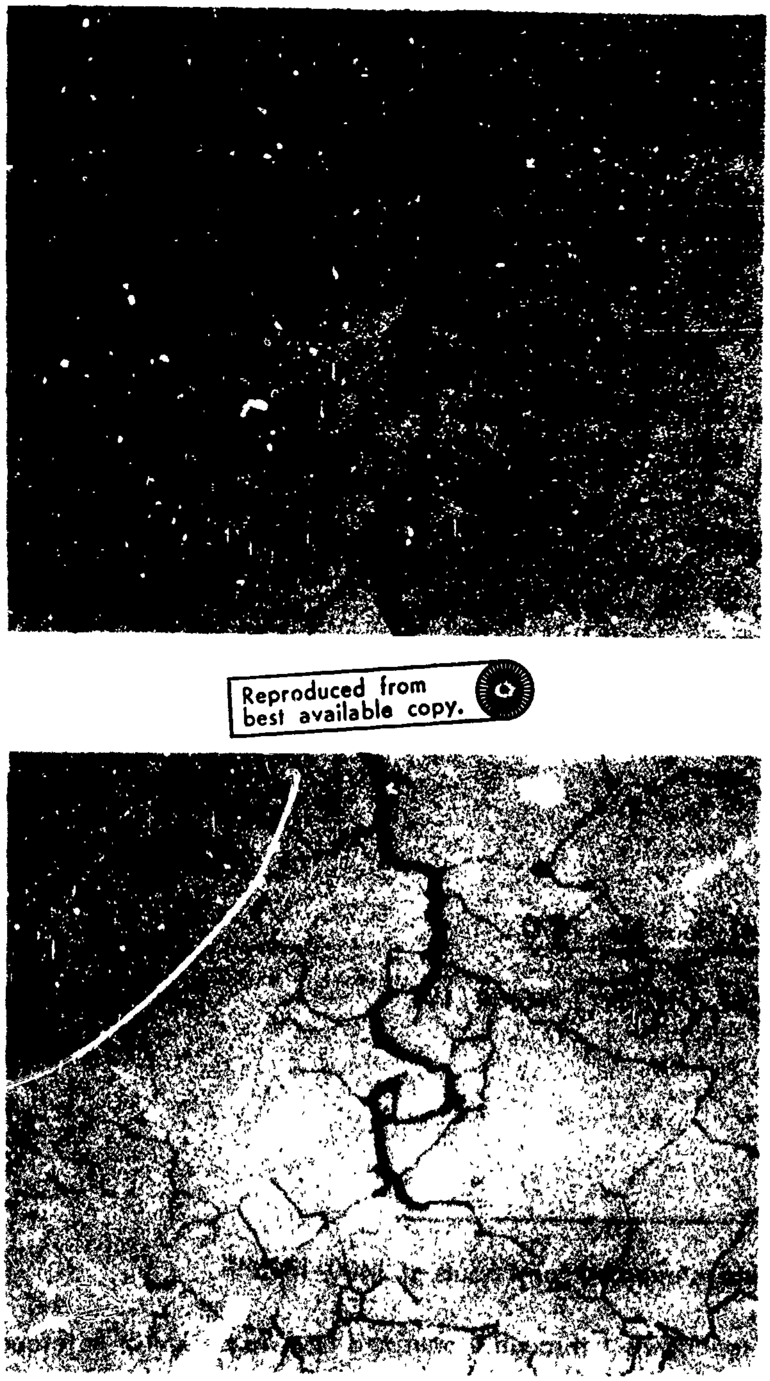

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FIGURE 19 - TYPICAL MICROSTRUCTURE ASSOCIATED WITH ISOTHERMAL HEAT-TREAT CRACKING IN REP POWDER SHEET SHOWING MICRO-DELAMINATION (TOP) AND CRACK-ARREST REGION (BOTTOM). ETCHANT: 60 HCl - 30 LACTIC - 7 HNO_3 - 3 H_2SO_4 - - X500

interparticle boundaries and/or rolled-out oxide networks. The lower photomicrograph of the crack extremity near the mid-thickness of the sheet shows the rather circuitous path of the crack along subboundary networks.

A possible contributing factor to cracking with REP isothermal tests was uncovered in a systematic metallographic study of some of the failed isothermal specimens. Several of these specimens were surface-ground to remove the weld reinforcement, sectioned, and then mounted for metallographic examination so that top-surface plane views of cracking networks could be examined. Representative photomicrographs of the microstructures along the weld-HAZ transition region as viewed from the top surface are shown in Fig. 20; the top micrograph shows a typical microcrack and the bottom one a crack-free region of the weld-HAZ transition structure. As indicated by four arrows in the top micrograph, there are pore-like areas associated with the crack. The formation of porosity at the fusion transition region is not an uncommon effect associated with the welding of powder metallurgy products such as tungsten and copper. This would suggest the possibility of crack formation by connection of closely spaced pores. The possibility that some of these pore-like sites are individual particles removed during metallographic preparation should not be ruled out. Despite the tendency towards cracking in the isothermal tests, it is quite apparent that the 1975°F-25 min. (WQ) solutioning treatment is not as effective in reducing the incidence of cracking in the REP sheet material as it is with the conventional Rene' 41 wrought sheet. The resistance to cracking of the pre-weld solutioned REP sheet specimens, however, is markedly better than that of the pre-weld mill annealed wrought sheet material.

Comparison of the isothermal crack-susceptibility data of the pre-weld solutioned basic heat (No. 7470) shown in Table VIII with the data reported in Ref. 1 for the pre-weld mill annealed condition also illustrates the effectiveness of the subject pre-weld solutioning treatment. No cracking was detected with the specimens noted in Table VIII, whereas gross cracking was experienced with 4 of the 5 pre-weld mill annealed condition specimens tested in the earlier program. In reviewing all of the crack-susceptibility data obtained with the REP sheet material, it can be

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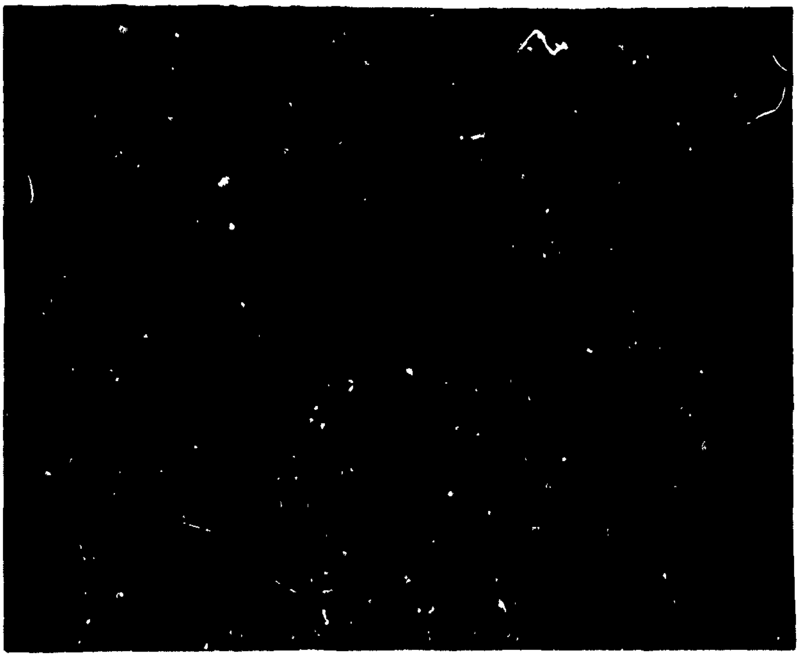

FIGURE 20 - SURFACE PLANE VIEW OF HEAT-TREAT CRACKING IN RENE' 41 REP POWDER SHEET ISOTHERMAL SPECIMENS. POROSITY ASSOCIATED WITH CRACKS (TOP) ALONG THE FUSION-HEAT AFFECTED ZONE INTERFACE (BOTTOM). ETCHANT: 60 HCl - 30 LACTIC - 7 HNO_3 - 3 H_2SO_4 - - X500

concluded that this powder metallurgy product is readily adaptable for welded fabrication and the material appears to exhibit considerable resistance to heat-treat cracking. The on-heating crack-susceptibility data obtained previously[1] indicated the ability of the REP sheet material to resist heat-treat cracking using a pre-weld solution treatment. These data are sufficiently encouraging for the consideration of this material in tension-critical applications, welded fabrication, and for forged components, such as blades, where weld repair (reclamation) would be necessary.

Procedure Verifications

The scope of the procedure verifications conducted in this program included (1) on-heating crack-susceptibility testing screening the applicability of pre-weld solution treatments for avoiding heat-treat cracking in various heats of Rene' 41 and Waspaloy and (2) on-heating hardening studies for the purpose of correlating hardening response with crack-susceptibility behavior. An additional purpose for the on-heating hardening tests was to explore the possibility of using hardness data instead of restraint test results to predict the heat-treat cracking sensitivity of these alloys. Ultimately, the tests could be used for incoming and production control of base metals to avoid the costly experience of producing hardware to determine whether an alloy or heat-treat condition is prone to heat-treat cracking.

Restraint Test Results

Results of verifications completed for two different base metals (Rene' 41 and Waspaloy) are summarized in Table IX. The data for Rene' 41, Heat No. 6842, were obtained for the pre-weld mill annealed condition and the solution annealed condition with two different post-solutioning quench rates. The quenching techniques are the same as those described earlier using Sen-Pak containers and the "CR-THA" heat-treat coating (see Tables II and III). The temperatures at onset of gross cracking for the pre-weld mill annealed specimens averaged 1340°F, whereas no gross cracking occurred with the pre-weld solutioned specimens. Tight HAZ microcracks similar to those shown in Fig. 8 were detected in all the specimens representing the slower post-solution quench rate condition. No microcracking was evident in the specimens fabricated

TABLE IX

ON-HEATING CRACK SUSCEPTIBILITY TEST RESULTS FOR VARIOUS "PROCEDURE VERIFICATIONS"

Specimen No.	Base Metal	Pre-Weld Base Metal Condition	Frame Alloy	Temp., Onset of Gross Cracking (°F)	Hardness[a] Pre-Weld	Hardness[a] After Test
1(1)	Rene' 41	Mill Annealed	AISI 304	1335		
2(2)	(Ht. No. 6842)		↓	1345		
3(5)	↓	"	↓	1340	30.0	40.0
4(35)	↓	1975°F-1/2 hr, WQ[c]	↓	NGC[g]		
5(36)	↓	"	↓	NGC[g]	28.0	39.0[f]
6(37)	↓	"	↓	NGC[g]		
7(83)	↓	1975°F-1/2 hr, WQ[d]	↓	NGC		
8(84)	↓	"	↓	NGC	R_A-57.5	39.5[f]
9(57)	Waspaloy	Mill Annealed	AISI 302	——[h]		
10(58)	(Ht. V2278)[b]	"	↓	1415	R_A-59.0	39.0
11(59)	↓	"	↓	1425		
12(60)	↓	"	↓	1475		
13(61)	↓	1975°F-1/2 hr, WQ[c]	↓	NGC		
14(62)	↓	"	↓	NGC	R_A-55.0	37.0[i]
15(63)	↓	"	↓	NGC		

Notes:
(a) Hardness values in unaffected base metals; R_C unless otherwise noted.
(b) Material obtained from D. S. Duvall, Pratt & Whitney Aircraft; material produced by Cannon Muskegon to Spec. PWA 1030D.
(c) Heat treated in Sen-Pak containers; a product of the Senrry Co., Foxboro, Mass.
(d) Heat treated with "CR-THA" coating; a product of the Markal Co., Chicago, Ill.
(e) NGC - No Gross Cracking.
(f) Includes 2-hr. dwell at 1400-1425°F.
(g) These specimens contained microcracks typical of those shown in Figure 3.
(h) Acoustic emission system difficulties; specimen did not exhibit gross cracking.
(i) Includes 2-hr. dwell at 1480-1500°F.

from the more rapid post-solution quenched material. The pre-weld hardness data for the Rene' 41 tests are consistent with cracking results. In comparison with the "basic heat" of Rene' 41 (Heat. No. 7470), the subject heat (No. 6842) has a somewhat higher mill annealed hardness and exhibits a lower gross cracking temperature (1340°F versus 1360°F). Both the slow (Sen-Pak) and rapid ("CR-THA") post-solutioned quenched pre-weld hardnesses are comparable to those obtained for the 1975°F solutioning treatments with the basic heat of Rene' 41 (refer to Tables II and III).

A change in frame alloy was required for the Waspaloy procedure verifications. One pre-weld mill annealed specimen (identified as #56) was fabricated with an AISI 304 stainless steel frame and on-heated to approximately 1500°F. This specimen did not undergo failure due to the inability to achieve a critical "thermal strain differential" between Waspaloy and the AISI 304 frame; the thermal expansivity of Waspaloy is greater than that of Rene' 41. Consequently, the subsequent specimens were fabricated with AISI 302 frames; the thermal expansivity of AISI 302 is greater than that of AISI 304. As shown in Table IX, all of the pre-weld mill annealed condition specimens experienced gross cracking with gross cracking temperatures ranging from 1415 to 1475°F. A direct correlation of the crack-susceptibility of Waspaloy and Rene' 41 is not conveniently possible with the subject tests due to their variations in thermal expansivities and the different expansivity values of the frame alloys. All of the pre-weld solution annealed condition specimens were found to be free of cracks despite the fact that the base metal was post-solution quenched in Sen-Pak containers (the slower quench rate technique). The resultant pre-weld hardness values for the solution annealed condition indicates that Waspaloy may not be as quench rate sensitive as Rene' 41. Resultant after-test hardnesses appear to be a little lower than those obtained with the heats of Rene' 41 studied in this program. Examples of gross cracking obtained with the pre-weld mill annealed condition Waspaloy tests are shown in Fig. 21. As with Rene' 41, Hastelloy W filler metal was used for welding the Waspaloy specimens.

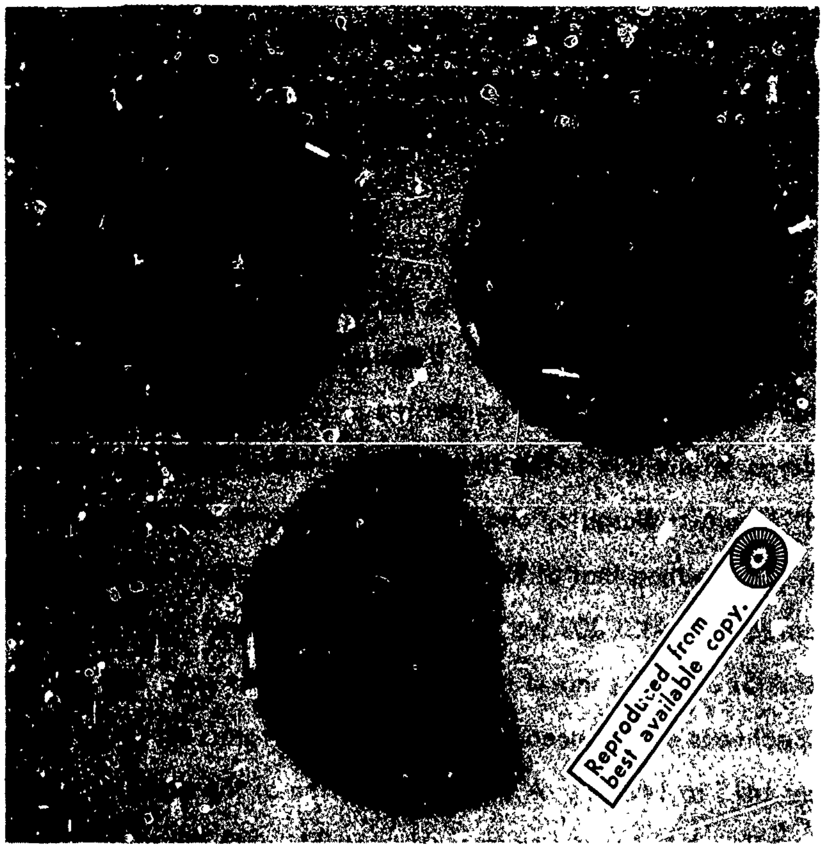

FIGURE 21 - GROSS CRACKING OF WASPALOY CRACK-SUSCEPTIBILITY SPECIMENS. BASE METAL WAS IN THE PRE-WELD MILL ANNEALED CONDITION.

To summarize, crack-susceptibility test results obtained for procedure verification purposes have shown that the 1975°F-1/2 hr (WQ) pre-weld solutioning treatment was effective for avoiding post-weld heat treat cracking with three heats of Rene' 41 and one heat of Waspaloy. All of these materials represented varying susceptibilities to heat-treat cracking in the mill annealed condition. The Rene' 41 materials included two production heats and one experimental heat[1]. One of the production heats (No. 6842) was an out-of-hardness grade and the experimental heat (No. 5939) was a low-carbon grade which was more prone to heat-treat cracking in the mill annealed than either of the production heats. The restraint test results obtained for the three heats of Rene' 41 and the one heat of Waspaloy are considered sufficient data to justify and recommend pre-weld solutioning as an approach to avoid heat-treat cracking with the intermediate hardener content nickel-base superalloys.

Another measure of the on-heating cracking resistance using the pre-weld 1975°F-1/2 hr (WQ) condition was demonstrated by substituting AISI 302 stainless for the AISI 304 on-heating specimen frames. These tests were conducted with the Rene' 41 basic heat (No. 7470). In order to obtain a more quantitative measure of the on-heating cracking resistance of the pre-weld 1975°F-1/2 hr (WQ) condition, two tests were conducted using AISI 302 stainless frames instead of the usual AISI 304 frame. The use of AISI 302 in place of AISI 304 increased the "thermal strain differential" during on-heating which increased the probability of obtaining some critical strain necessary for crack initiation and failure. As the results in Table X indicate, gross cracking was encountered during on-heating, however, failure occurred in the weld metal of the outer frame-to-disk weld. These results, although not anticipated, indicate that the 1975°F-1/2 hr (WQ) pre-weld treatment imparts sufficient crack resistance or strain (relaxation) capacity to withstand substantial stresses during on-heating. As a point of reference, as noted in Table X, gross cracking was encountered at 1340°F with the pre-weld mill annealed condition using AISI 302 frames and 1360°F with the pre-weld annealed condition using AISI 304 frames.

TABLE X

"ON-HEATING" CRACK-SUSCEPTIBILITY TEST RESULTS FOR RENE' 41 (HT. NO. 7470) SHEET WELDMENTS

Specimen No.	Pre-Weld Base Metal Condition	Temp., Onset of Gross Cracking (°F)	Frame Alloy
1(32)	1975°F-1/2 hr (WQ)	1360[a]	AISI 302 Stainless
2(33)	"	1365[a]	"
Reference[1]	1975°F-1/2 hr (WQ)	No Gross Cracking	AISI 304 Stainless
"	Mill Annealed	1360	"
"	"	1340	AISI 302 Stainless

Notes: (a) Weld metal failures in disk to frame (outer) welds; cracking was circumferential along the centerline of the welds.

On-Heating Hardness Evaluation

As noted earlier, the primary purposes for obtaining these data are to (1) correlate on-heating hardening response for crack-susceptible and crack resistance conditions for each material used for verification purposes and (2) determine whether a correlation exists between the hardening characteristics of the various materials and their susceptibility to heat-treat cracking.

On-heating hardness data obtained for two Rene' 41 heats (No. 5939 and 6892) and one heat of Waspaloy are plotted in Figs. 22, 23, and 24; these should be compared with the on-heating hardness curves for the Rene' 41 basic heat (No. 7470) shown in

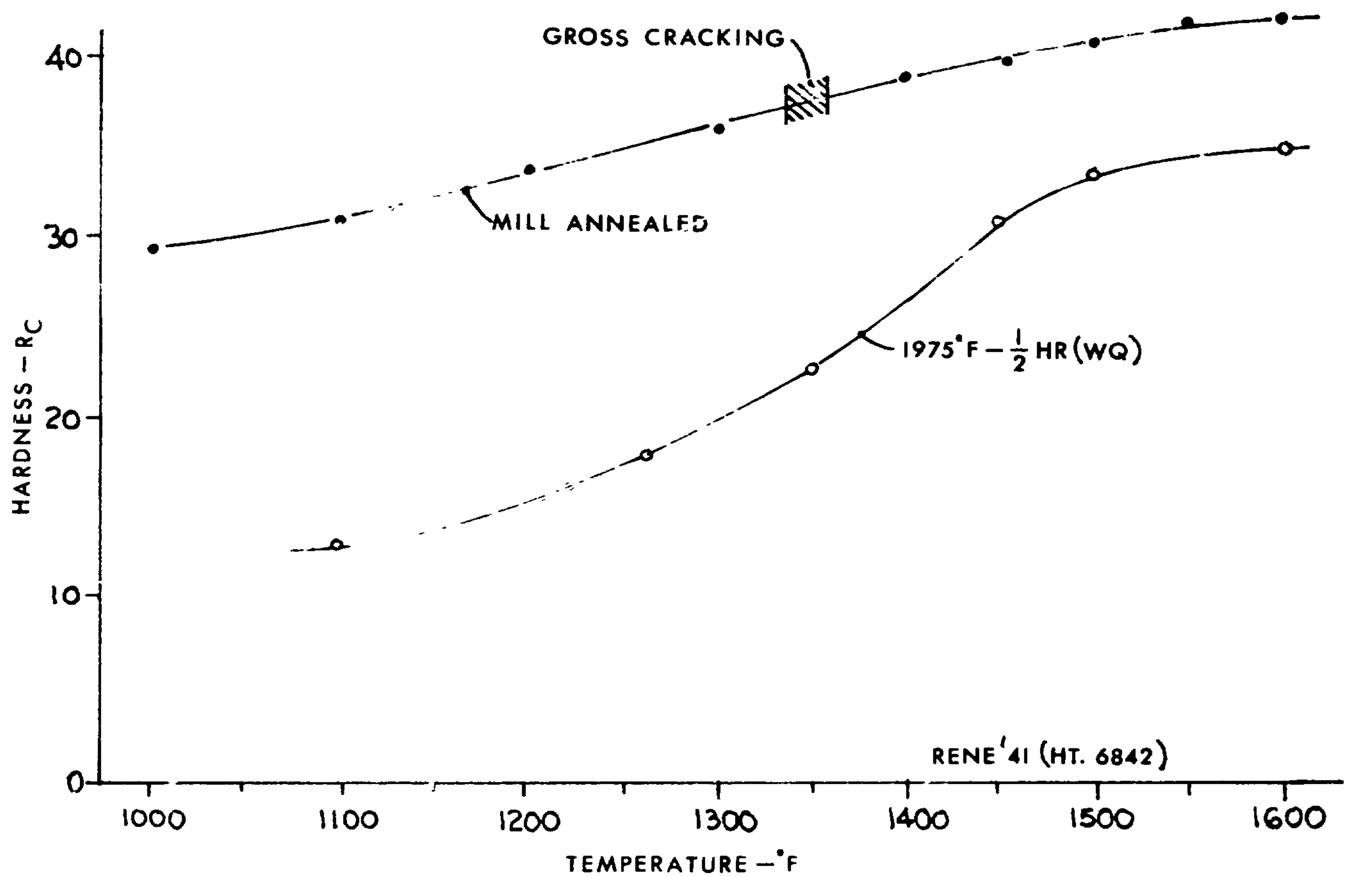

FIGURE 22 - ON-HEATING HARDENING RESPONSE OF RENE' 41 (HT. 6832) FOR THE MILL ANNEALED AND SOLUTIONED (1975°F-1/2 HR, WQ) CONDITIONS.

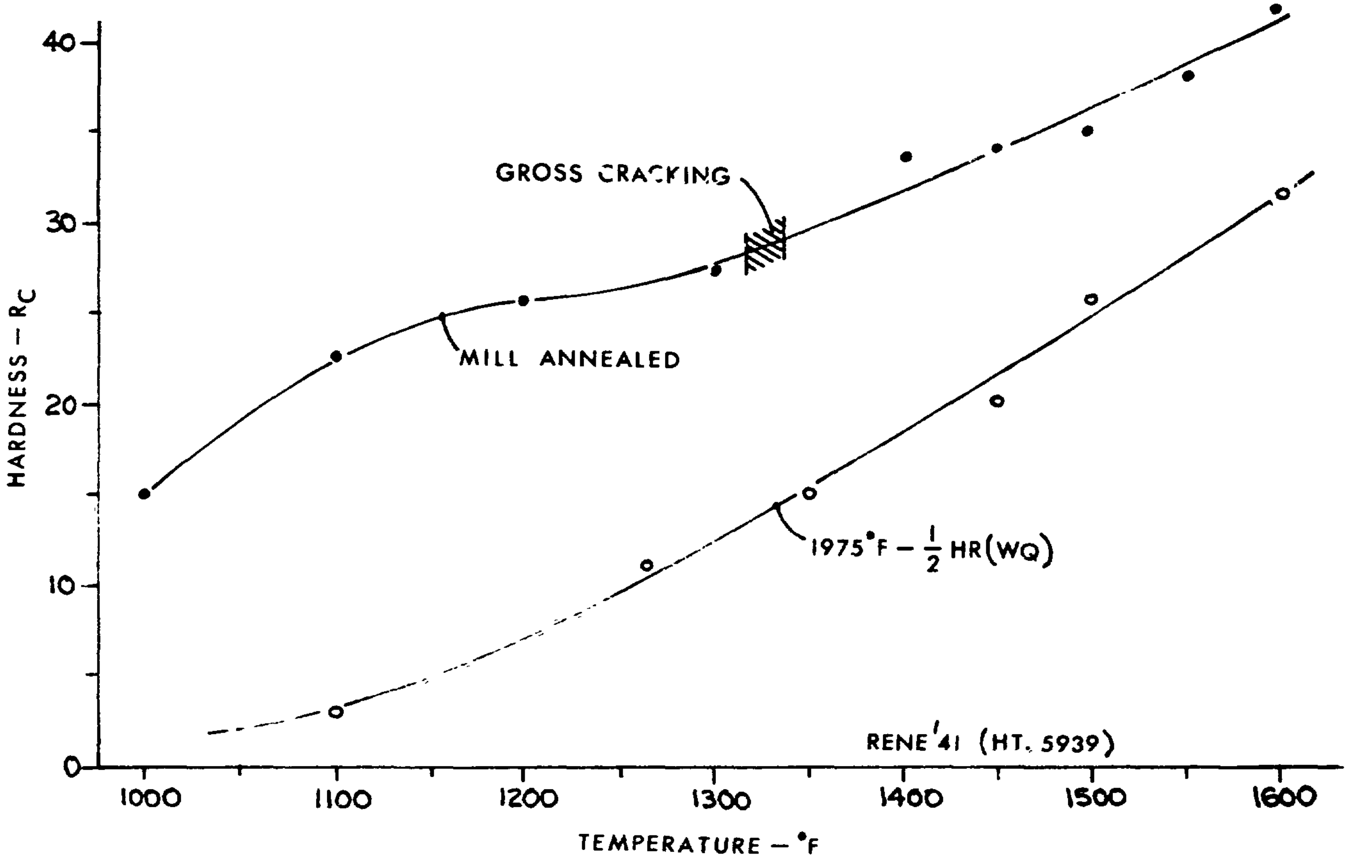

FIGURE 23 - ON-HEATING HARDENING RESPONSE OF RENE' 41 (HT. 5939) FOR THE MILL ANNEALED AND SOLUTIONED (1975°F-1/2 HR, WQ) CONDITIONS.

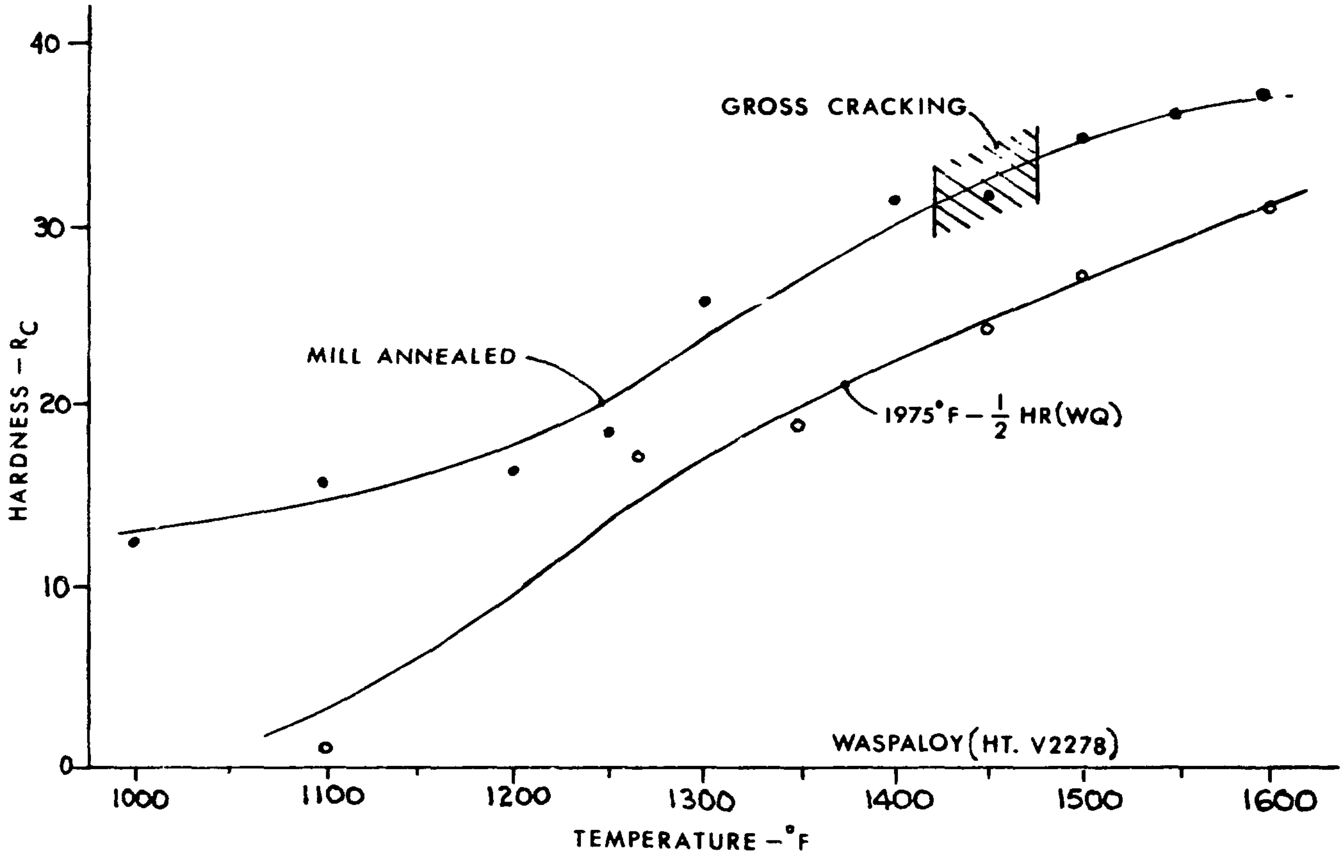

FIGURE 24 - ON-HEATING HARDENING RESPONSE OF WASPALOY (HT. V2278) FOR THE MILL ANNEALED AND SOLUTIONED (1975°F-1/2 HR, WQ) CONDITIONS.

Fig. 9. These curves are for the mill annealed and solutioned (1975°F-1/2 hr, WQ) conditions with the gross cracking temperature range indicated on the mill annealed curves in each instance. In each case, a "slack-hardening" response is noted with the most pronounced effect observed with Rene' 41 Ht. No. 6842 and the least effect with the Waspaloy heat. The reasons for these variations are obvious; the Rene' 41 Heat No. 6842 material was in a high-hardness mill annealed condition, whereas the Waspaloy heat was a low mill annealed hardness condition.

A comparison of the on-heating curves of Figs. 9, 22, 23, and 24, does not show a clear correlation between crack-susceptible and crack resistance conditions; however, there are several aspects of the curves which should be noted. If the hardness (indirectly, strength level) at the temperature of gross cracking for the mill annealed conditions were considered critical, then the comparable hardness associated with the solutioned condition would be of interest. This comparable hardness on the solution annealed curves for all of these materials is attained at 1550°F or higher during on-heating, which is sufficient in both temperature and time to allow considerable stress relaxation to take place. That is, the critical hardness is obtained after restraint stresses are lowered by straining of the bulk (unaffected) base metal and portions of the HAZ. Of primary interest in this phase of the program was the possible isolation of conditions and/or parameters which could be used for specification requirement purposes. One possibility suggested by the on-heating data discussed here is a requirement that the hardness of the base metal should not exceed 30 R_C when on-heated to 1400°F at the rate of 15-17°F/minute. The data presented here are not considered sufficient to rigorously support this requirement and it is strongly recommended that more on-heating data be developed with sufficiently varied heats of material for this purpose.

Astroloy Pre-Weld Heat Treatments

One of the major objectives of this program was to develop procedures for avoiding cracking in higher hardener content superalloys which, to date, have been generally considered unweldable. Astroloy sheet material was procured for this purpose; some

earlier preliminary results using this material are reported in Ref. 1. It was found in this earlier work that a simple pre-weld solutioning treatment, as successfully used with Rene' 41, was effective in improving the resistance of Astroloy to on-heating cracking. However, due to the rapid hardening response attributable to a high hardener content, pre-weld solutioning did not appear to be a likely approach for avoiding heat-treat cracking in this alloy. Consequently, the current work was directed to evaluating pre-weld heat treatments which produce a less metastable microstructure than that obtained by solution treatment. The primary objective was to achieve a structure with a minimum pre-weld hardness and a sluggish response to hardening during the on-heating interval. This would permit stress relaxation to occur earlier during the on-heating interval before attaining a limiting strain due to precipitation hardening.

Effects of Pre-Weld Solutioning

As with the Rene' 41 pre-weld solutioning tests, the effects of solutioning time and temperature were also investigated for Astroloy. On-heating crack-susceptibility data obtained in the preceding program(1) showed that the "gross cracking temperature" was increased from about 1180°F for the mill annealed condition to approximately 1295°F using the 1975°F-1/2 hr (WQ) pre-weld solution treatment. The effects of varying solutioning parameters on heat-treat crack susceptibility are summarized in Table XI. As noted, only three specimens of this series were successfully fabricated without experiencing cracking during welding. The remainder of the specimens cracked during the cooling interval following completion of the inner (test) weld. All cracking was associated with the inner weld and was similar to the on-heating cracking obtained with the pre-weld mill annealed condition in the previous program(1). Examples of the noted cracking for two of these specimens are shown in Fig. 25. The gross cracking temperatures indicated for the three specimens tested are all somewhat lower than those obtained previously(1) for the 1975°F-25 min (WQ) treatment suggesting that prolonged solutioning times are not beneficial with respect to improving crack resistance. The data in Table XI also indicate that increasing temperatures within the solurioning temperature range increases the

TABLE XI

"ON-HEATING" CRACK-SUSCEPTIBILITY DATA SHOWING EFFECTS OF PRE-WELD SOLUTIONING TREATMENTS ON THE SUSCEPTIBILITY OF ASTROLOY (HT. NO. 6250) WELDMENTS TO HEAT-TREAT CRACKING

Specimen No.	Pre-Weld Base Metal Condition	Temp., Onset of Gross Cracking (°F)	Hardness[a] Before Test	Hardness[a] After Test
1(30)	1975°F - 1 hr., WQ	1270	43.5	43.0
2(31)	"	1240		
3(34)	1975°F - 4 hr., WQ	Did not test[b]	41.0	--
4(38)	"	"		
5(14)	2050°F - 1/2 hr., WQ	1240	40.0	42.0
6(15)	"	Did not test[b]		
7(16)	"	"		
8(25)	2175°F - 1/2 hr., WQ	Did not test[b]	40.0	--

Notes: (a) Rockwell "C" hardness values taken in unaffected base metal.

(b) Specimens cracked during fabrication; see Fig. 25 of typical cracking.

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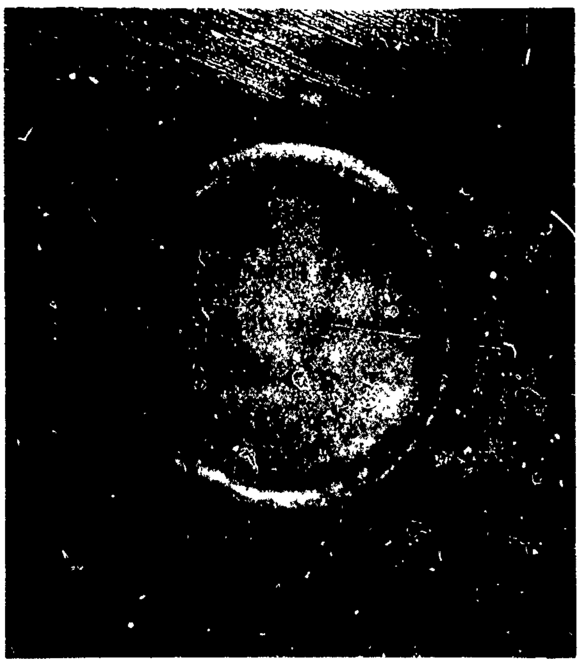

FIGURE 25 - RESTRAINT CRACKING OF ASTROLOY SPECIMENS WHICH OCCURRED DURING FABRICATION. TOP: SPECIMEN 4(38) - X5; BOTTOM: SPECIMEN 8(25) - X1.25

susceptibility to cracking during welding and post-weld thermal cycling. One consideration which would have an influence on the results shown in Table XI is the effectiveness of the post-solution quench. These specimens were fabricated from base metal which was solution heat treated in "Sen-Pak" containers, whereas the specimens of the previous effort[1] were fabricated from base metal heat treated with the "CR-THA" coating. Consequently, the pre-weld base metal hardnesses noted in Table XI, are approximately ten Rockwell C points higher than those obtained in the prior work. The hardness variations between these quench techniques are more pronounced with Astroloy than with Rene' 41 due to its more rapid aging response.

Effects of Duplex Solutioning-Overaging Treatments

For reasons noted previously regarding the use of on-heating hardness data, considerable effort was spent studying the on-heating hardening response of Astroloy. Some of the initial pre-weld heat treatments selected for Astroloy are summarized in Table XII along with respective hardness data. The purpose of screening the solutioning treatments shown was to establish some minimum hardness levels that might be expected for this alloy. The use of a coating material to achieve a greater post-solution quench rate is also reflected for two of the three solutioning treatments shown in the table. As these data indicate, the apparent minimum hardness that might be expected with this alloy is about 31.0 R_C. Hardnesses obtained with larger pieces of material for the same solutioning treatments were three to four points higher, indicating further a quite rapid response to hardening just due to mass differences. Crack-susceptibility tests were not conducted using these pre-weld solutioning treatments for reasons cited earlier. Comparing the 2160°F and the 1975°F-1/2 hr treatment suggests that a slightly lower hardness is possible with a lower γ' solutioning temperature.

The first two pre-weld "overaging" treatments listed in Table XII are in reality precipitation-conditioning treatments. As the data indicate, these had little affect with respect to lowering hardness; the mill annealed condition hardness for this material was 41.0 R_C. The reasoning for selecting the third overaging treatment listed in Table XII is as follows: the 16 hr-1975°F (WQ) cycle would produce an effectively homogenized structure in which coarse γ' would be rapidly produced at

TABLE XII

SUMMARY OF AS-HEAT TREATED HARDNESSES FOR VARIOUS EXPLORATORY PRE-WELD SOLUTIONING AND OVERAGING TREATMENTS FOR ASTROLOY SHEET MATERIAL

Pre-Weld Treatment[a]	Hardness (R_c)	Remarks
Solutioning Treatments:		
1975°F-1/2 hr. (WQ)	31.5	
1975°F-16 hrs. (WQ)	31.5	
1975°F-16 hrs. (WQ)	30.5	Coated[b]
2160°F-1/2 hr. (WQ)	34.5	
2160°F-1/2 hr. (WQ)	34.0	Coated[b]
Overaging Treatments:		
MA + 1650°F-4 hrs. (FC) + 1850°F-16 hrs. (WQ)	40.5	
1975°F-1/2 hr. (WQ) + 1650°F-4 hrs. (FC) + 1850°F-16 hrs. (WQ)	38.0	
1975°F-16 hrs. (WQ) + 1850°F-24 hrs.	31.5	
1975°F-2 hrs. (WQ) + 1850°F-X hrs.	See Fig. 26	
2050°F-4 hrs. (WQ) + 1850°F-X hrs.	See Fig. 26	

Notes: (a) WQ - water quenched; FC - furnace cooled to 1850°F.

(b) These specimens were coated with "CR-THA" Protective Coating, a product of the Markal Co., Chicago, Illinois, to provide a more rapid quench rate than that obtained with water quenching.

1850°F and further agglomerated during the prolonged 24-hour exposure. As the data indicate, this treatment produces a structure with a hardness comparable to that obtained by simple solution treatment.

Two other similar overaging treatments were also selected as shown in Table XII using slightly different solutioning cycles. The hardness data as a function of aging time at 1850°F for these two pre-weld treatments are plotted in Fig. 26. The significance of these data is that the terminal hardness is influenced by the slightly different solution treatments used. In this case, there is approximately a two-point hardness difference and the higher solution temperature is associated with the higher hardness values. It is noteworthy that the minimum hardness shelf is approached within 20 to 25 hours with both solutioning treatments.

Since the hardness results for the 1975°F-2 hr (WQ) solutioning treatment appeared to be consistent with those obtained with the 1975°F-16 hr (WQ) treatment noted in Table XII, on-heating hardness data were obtained for the two conditions. These data are shown in Fig. 27 along with the on-heating hardness curves for the mill annealed and solutioned (1975°F-1/2 hr (WQ) conditions for which crack-susceptibility results were obtained earlier. The data show that a significant variation in hardening response occurs between the 1975°F-2 hr and 1975°F-16 hr pre-weld heat-treat conditions. The response to hardening of the 2 hr solutioned material occurs early in-time and approaches hardness level comparable to those obtained by the as-solutioned (1975°F-1/2 hr, WQ) material which experienced gross cracking in earlier tests. For this reason, the 1975°F-2 hr treatment did not appear to be a pre-weld treatment to select for crack-susceptibility evaluation. By contrast, the 1975°F-16 hr is effective in producing a delayed response to hardening during on-heating. As the data of Fig. 27 indicate, hardening is delayed until the 1450-1500°F temperature range after which the rate of hardening is quite rapid. This particular pre-weld heat treatment appeared to be promising since the delay in hardening would allow stress relaxation to take place during on-heating and improve the chances of avoiding heat-treat cracking.

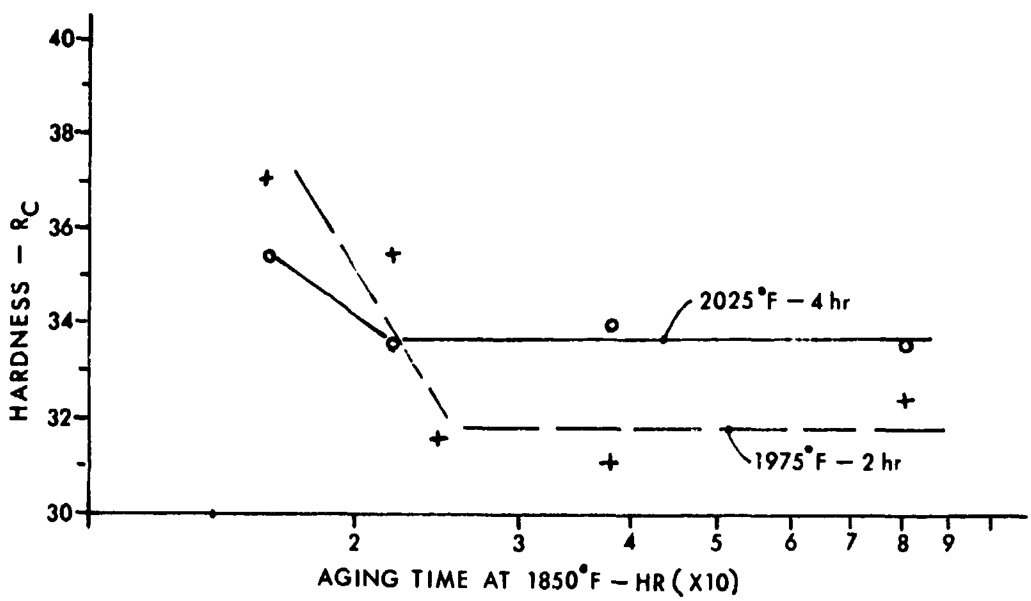

FIGURE 26 - ASTROLOY HARDNESS VARIATIONS AS A FUNCTION OF AGING TIME AT 1850°F FOR TWO PRE-AGE SOLUTION TREATMENTS.

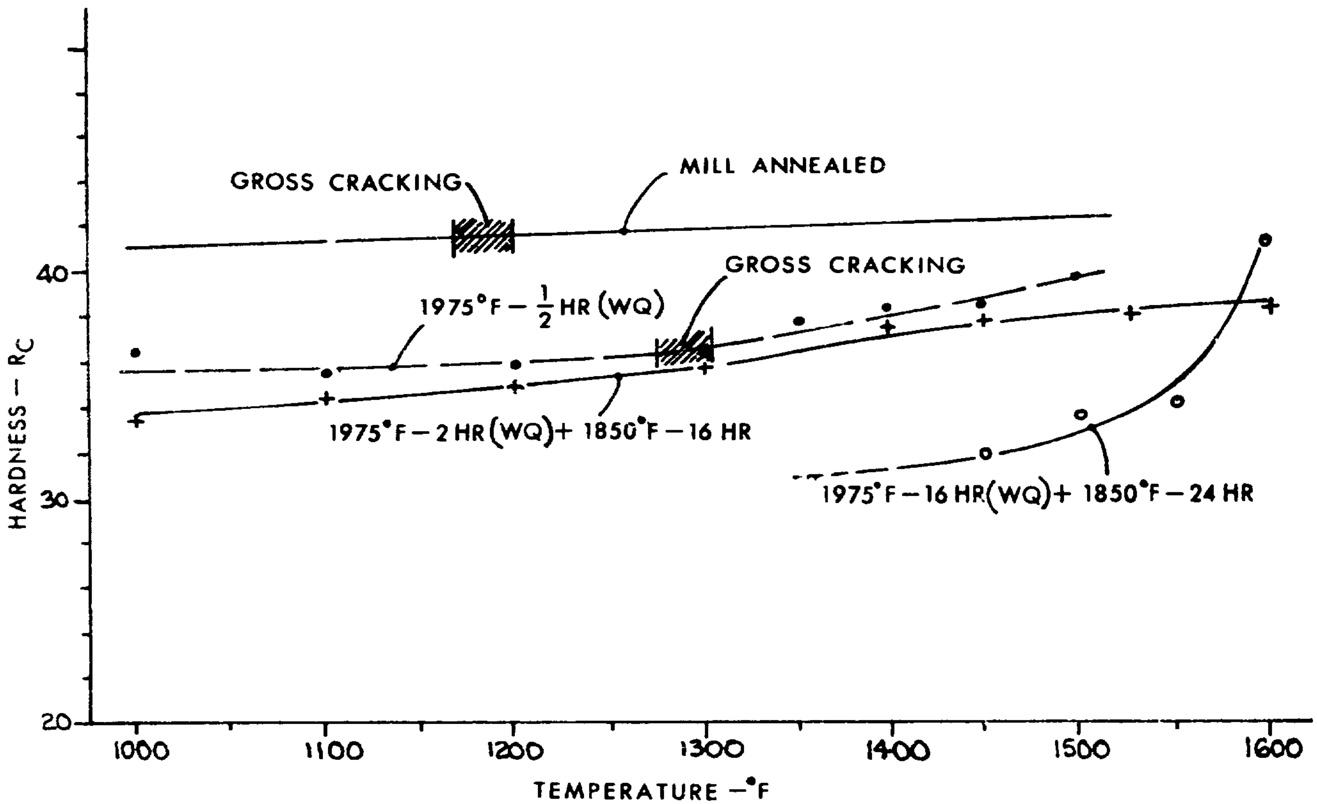

FIGURE 27 - ON-HEATING HARDENING RESPONSE OF ASTROLOY FOR VARIOUS PRE-WELD HEAT TREAT CONDITIONS.

Results of recent work by Duvall and Owczarski[15] has shown that a "2140°F-4 hr (forced air cooled) plus 1975°F-16 hr (cool to 1850°F at 100°F/hr) plus 1850°F-4 hr (cool 50°F/hr to 1650°F, 100°F/hr to 1050°F, cool to R.T.)" pre-weld heat treatment improves the resistance of Udimet 700 (similar to Astroloy) to heat-treat cracking. Consequently, Astroloy on-heating hardening data were obtained for this heat treatment along with another Lockheed-developed overaging pre-weld treatment. These data are plotted in Fig. 28. Both pre-weld treatments exhibit some metastability during on-heating with the Lockheed heat treatment exhibiting slightly lower hardness values up to about 1550°F. The P & W overage exhibits a slight softening effect at around 1500°F which would appear to be quite desirable for improving heat-treat crack resistance. In both cases, the heat treatments appear to satisfy the conditions of a low pre-weld hardness and a minimum hardening response during on-heating. The 1600°F hardness levels for these pre-weld treatments are considerably lower than that shown by the most favorable pre-weld heat treatment in Fig. 27.

Since the overaged Astroloy microstructures were found to be metastable during the on-heating interval, it was considered worthwhile to obtain on-heating hardness data for overaged Rene' 41 using the pre-weld heat treatment developed by Hughes and Berry in a previous AFML-sponsored program[9]. The data obtained for three heats of Rene' 41 and one heat of Waspaloy using this overaging treatment are plotted in Fig. 29. All of these heats were metastable during on-heating and exhibited a softening effect at about 1500°F. The low carbon Rene' 41 (Ht. 5939) and the Waspaloy material showed the most marked softening effect beginning at about 1400°F. There was essentially no difference in the on-heating responses of Rene' 41 Ht. No. 6842 and 7470 which may be attributed to their quite similar chemical compositions. The low carbon Rene' 41 material showed a slight increase in hardness prior to the softening trend, whereas the others did not vary in hardness until the softening effect occurred. The data of Figs. 28 and 29 are considered significant in that they indicate the difficulties of developing a truly stable microstructure for these materials. It appears that the past and presently defined overaging treatments are not really overaging treatments in the normal sense, and that the γ' precipitation kinetics of this alloy system cannot be suppressed throughout

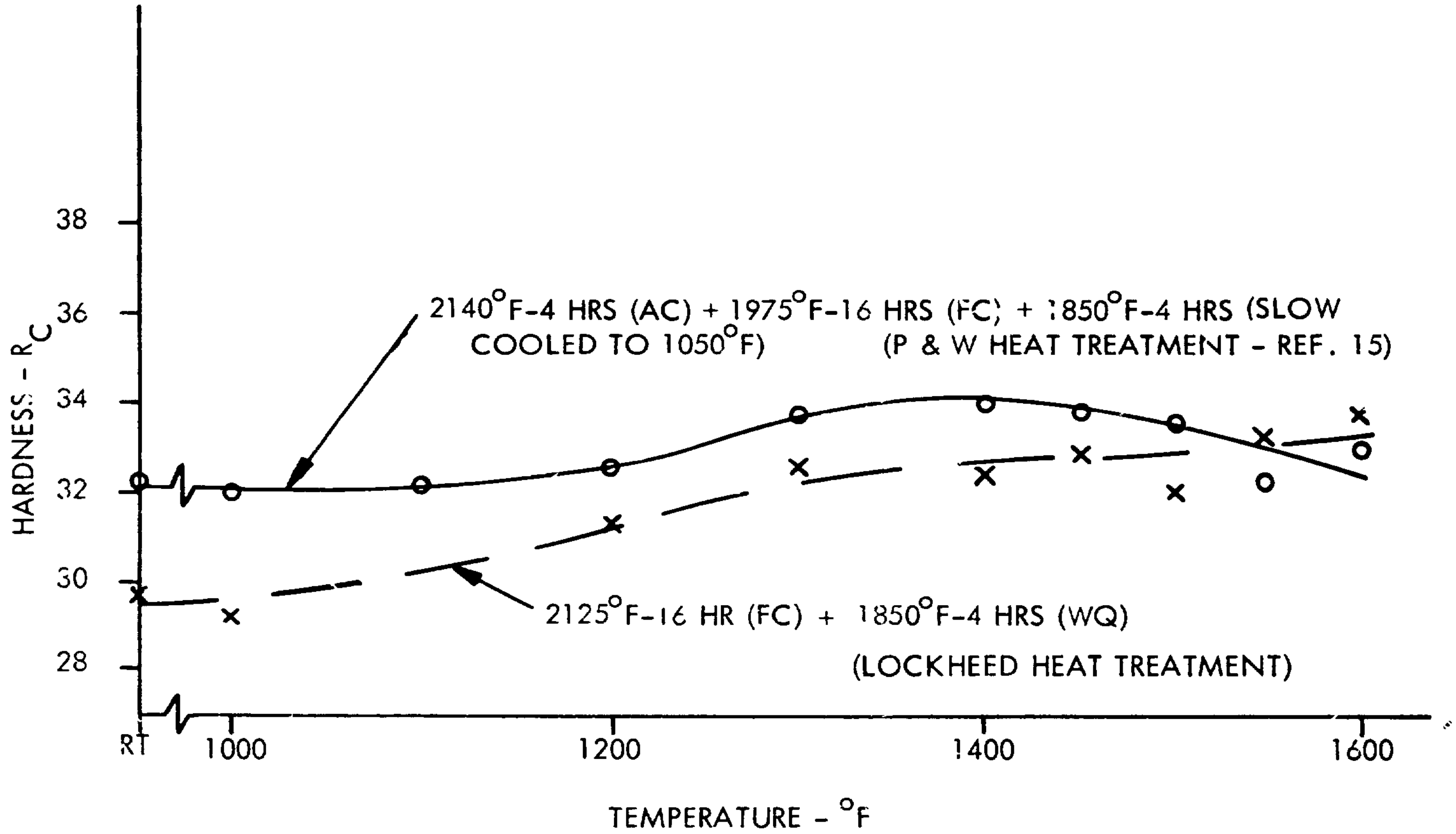

FIGURE 28 - ON-HEATING HARDNESS DATA FOR TWO ASTROLOY OVERAGING TREATMENTS.

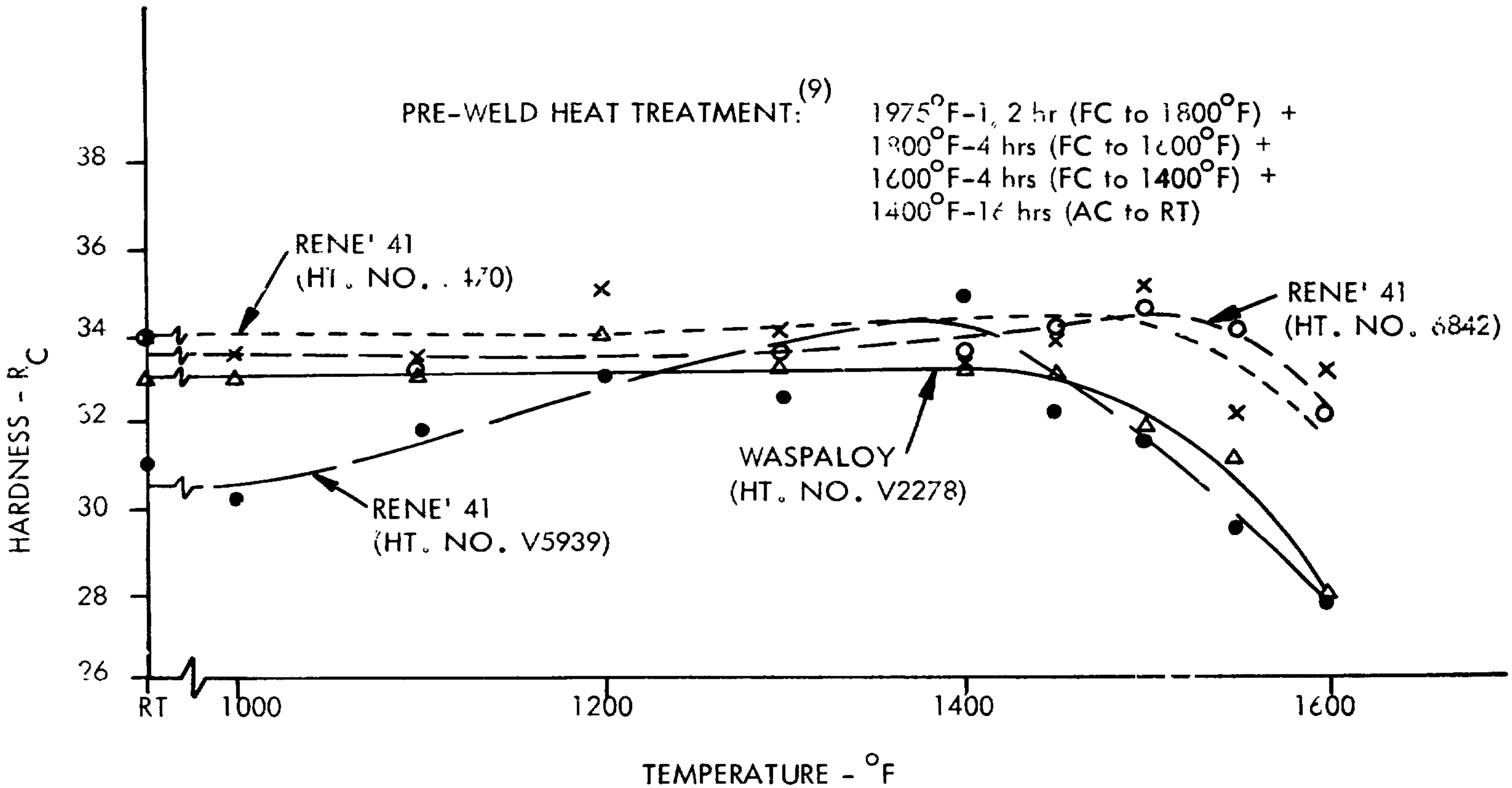

FIGURE 29 - ON-HEATING HARDNESS DATA FOR THE OVERAGED CONDITION IN THREE HEATS OF RENE' 41 AND ONE HEAT OF WASPALOY.

the aging temperature range. It is interesting to note that the hardness levels (see Figs. 28 and 29) associated with overaged Astroloy, Rene' 41, and Waspaloy structures are not too varied despite the hardener content differences.

Crack-Susceptibility Evaluations

Some initial on-heating crack-susceptibility data obtained for various Astroloy pre-weld heat treatments are summarized in Table XIII; these data were obtained before it was decided to use on-heating hardness evaluations for heat treatment screening purposes. The pre-weld treatments noted in this table were selected as possible simplified overaging treatments consisting of reasonably short γ' solutioning periods followed by high-temperature aging during which course γ' particles would be obtained. In comparing the data with the gross cracking temperatures for the pre-weld solutioned condition obtained previously, these pre-weld heat treatments do not appear to show much promise for significantly improving the resistance to heat-treat cracking. The latter four specimens were fabricated with Inconel 600 frames as an approach to lowering the "thermal strain differential" and possibly providing a somewhat more sensitive indicator by increasing the gross cracking temperatures. The change in frame alloys did increase the gross cracking temperatures as noted by comparing specimen 1(86) with 3(94) and 4(93), however, the increase was not considered sufficient to improve the sensitivity of the test. The lower gross cracking temperatures of specimens 5(92) and 6(91) clearly show the deleterious effects of repair welding. The gross cracking temperature of specimen 2(88) suggests that the foreshortened "1975^{o}F-1/2 hr (WQ) + 1825^{o}F-6 hrs (AC)" treatment might have been worthy of further study.

Crack-susceptibility data screening the effects of the two pre-weld overaging treatments noted in Fig. 28 are summarized in Table XIV. As the data show, all specimens fabricated with material in these heat-treat conditions failed during post-weld cooling and were not tested. The primary significance of these results and possibly of those shown in Table XI, is that prolonged high temperature pre-weld heat treatments increase the susceptibility of Astroloy to weld restraint cracking. This may be simply the result of grain growth rather than precipitation behavior. Cracking during welding

TABLE XIII

"ON-HEATING" CRACK SUSCEPTIBILITY DATA SHOWING EFFECTS OF SELECTED PRE-WELD DUPLEX HEAT TREATMENTS ON THE SUSCEPTIBILITY OF ASTROLOY (HT. NO. 6250) TO HEAT-TREAT CRACKING

Specimen No.	Pre-Weld Heat Treatment[a]	Temp., Onset of Gross Cracking (°F)	Frame Alloy
Reference[b]	Mill Annealed	1180	AISI 304
"	1975°F-1/2 hr (WQ)	1295	"
1(86)	1975°F-4 hrs (FC) + 1825°F-16 hrs (AC)	1230	AISI 304
2(88)	1975°F-1/2 hr (WQ) + 1825°F-6 hrs (AC)	1345	"
3(94)	1975°F-4 hrs (FC) + 1825°F-16 hrs (AC)	1290	Inconel 600
4(93)	"	1345	"
5(92)	"	1210[c]	"
6(91)	"	1230[c]	"

Notes: (a) Coding: WQ - water quench, FC - furnace cooled to 1825°F, AC - air cooled to room temperature

(b) Technical Report AFML-TR-70-24, November 1970

(c) Approximately 30° of inner weld was rewelded to remove crater cracks

TABLE XIV

ON-HEATING SPECIMEN FABRICATION DATA FOR PRE-WELD OVERAGED ASTROLOY (HT. NO. 6250)

Specimen No.	Pre-Weld Heat Treatment[a]	Frame Alloy	Remarks
1(112)	2125°F-16 hrs (FC) + 1850°F-4 hrs (WQ)	Inconel 600	All specimens failed during post-weld cooling. Failure was similar to heat-treat cracking - circumferentially along outside of inner weld in the immediate HAZ.
2(113)	"	"	
3(114)	"	"	
4(118)	"	"	
5(115)	2140°F-4 hrs (AC) + 1975°F-16 hrs (FC) + 1850°F - 4 hrs (Cool 50°F/hr to 1650°F, 100°F/hr to 1950°F, cool to RT)[b]	Inconel 600	
6(116)	"	"	
7(117)	"	"	

Notes: (a) Coding: FC - furnace cool to 1850°F, WQ - water quench

(b) Pre-Weld heat treatment developed by Duvall and Owczarski - see Ref. 15

could be avoided by modifying the specimen design to reduce restraint stresses. However, since specimens fabricated with material in various other pre-weld conditions were successfully welded without cracking and subsequently heat treated, it did not appear logical to reduce restraint conditions to obtain data for pre-weld treatments which did not look promising. Correlation of the pre-weld overaging data of Table XIV and Fig. 28 indicates that the use (or validity) of on-heating hardening evaluations for material control purposes must be substantiated with crack-susceptibility tests. That is, on-heating data such as shown in Fig. 28 are not alone sufficient to predict the relative crack-susceptibility of a pre-weld condition. Once a correlation is established between on-heating response and the cracking behavior associated with a given pre-weld condition, on-heating hardening evaluations may be confidently used for material acceptance and production control purposes.

Examination of specimen failures from tests reported in Table XIV revealed a grain (rolling) direction influence similar to that observed with Rene' 41[1]. Several of these failures are shown in Fig. 30. These examples clearly illustrate that heat-treat cracking is strongly influenced by grain direction although the specific role of grain orientation on crack initiation events is not readily apparent. The primary value of this observation is that grain direction should be considered in the design of Astroloy weldments and as an integral part of any heat-treat cracking control plan used for the nickel-base superalloys. The maximum or principal stress direction should be parallel with the grain orientation of the sheet to minimize the influence of the transverse orientation on heat-treat cracking.

Astroloy Microstructure Studies

An extensive metallographic study was conducted in conjunction with the Astroloy on-heating and crack-susceptibility evaluations for the purposes of characterizing microstructures and establishing a correlation between microstructure detail and cracking behavior. These studies included optical, replica, and scanning electron microscopy analyses for most of the Astroloy pre-weld heat treatments evaluated in the program. Representative microstructures for varying pre-weld base metal conditions are summarized in Figs. 31 through 36.

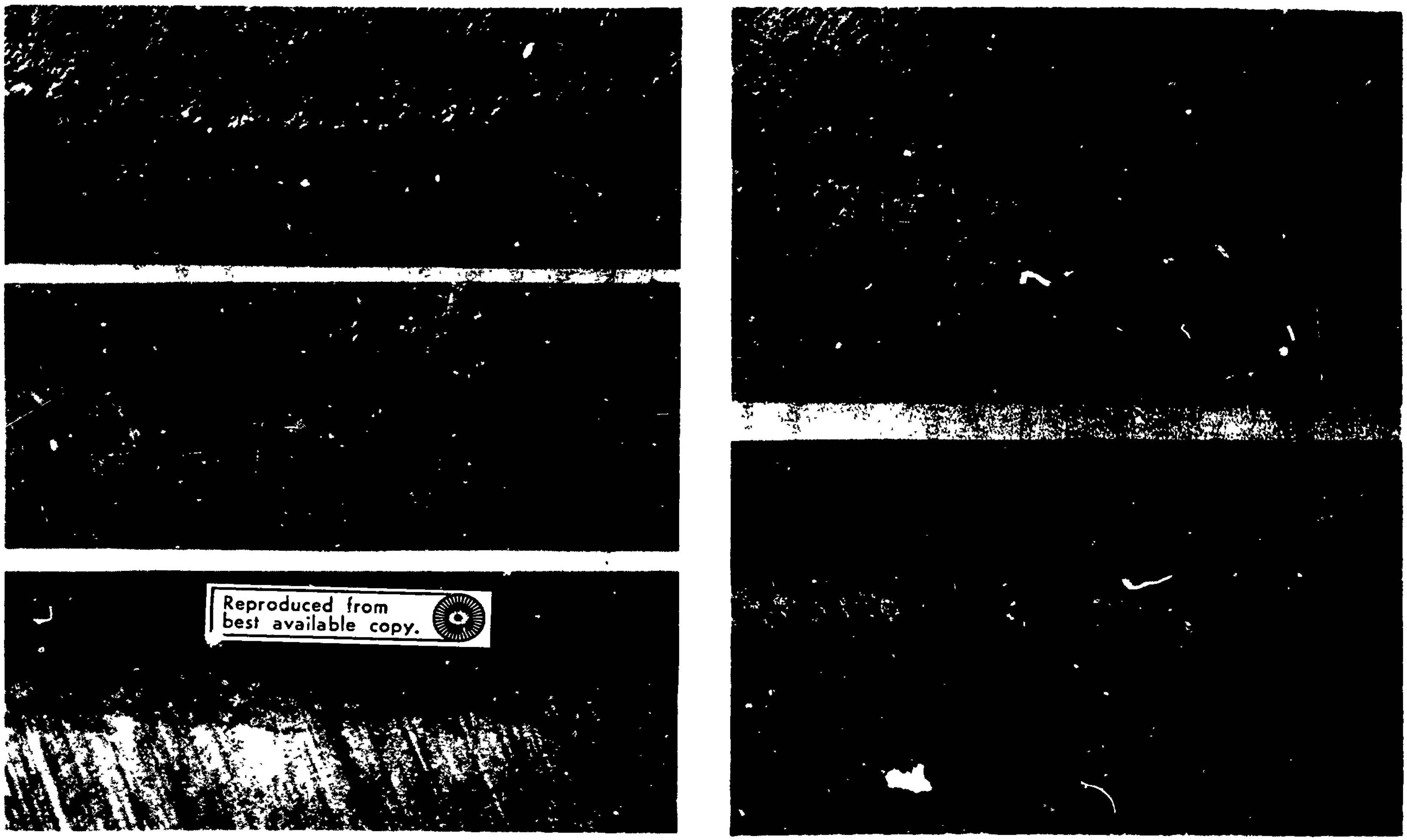
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FIGURE 30 - HEAT AFFECTED-ZONE CRACKING IN ASTROLOY CRACK-SUSCEPTIBILITY SPECIMENS SHOWING THE INFLUENCE OF GRAIN (ROLLING) DIRECTION. Appx. X12

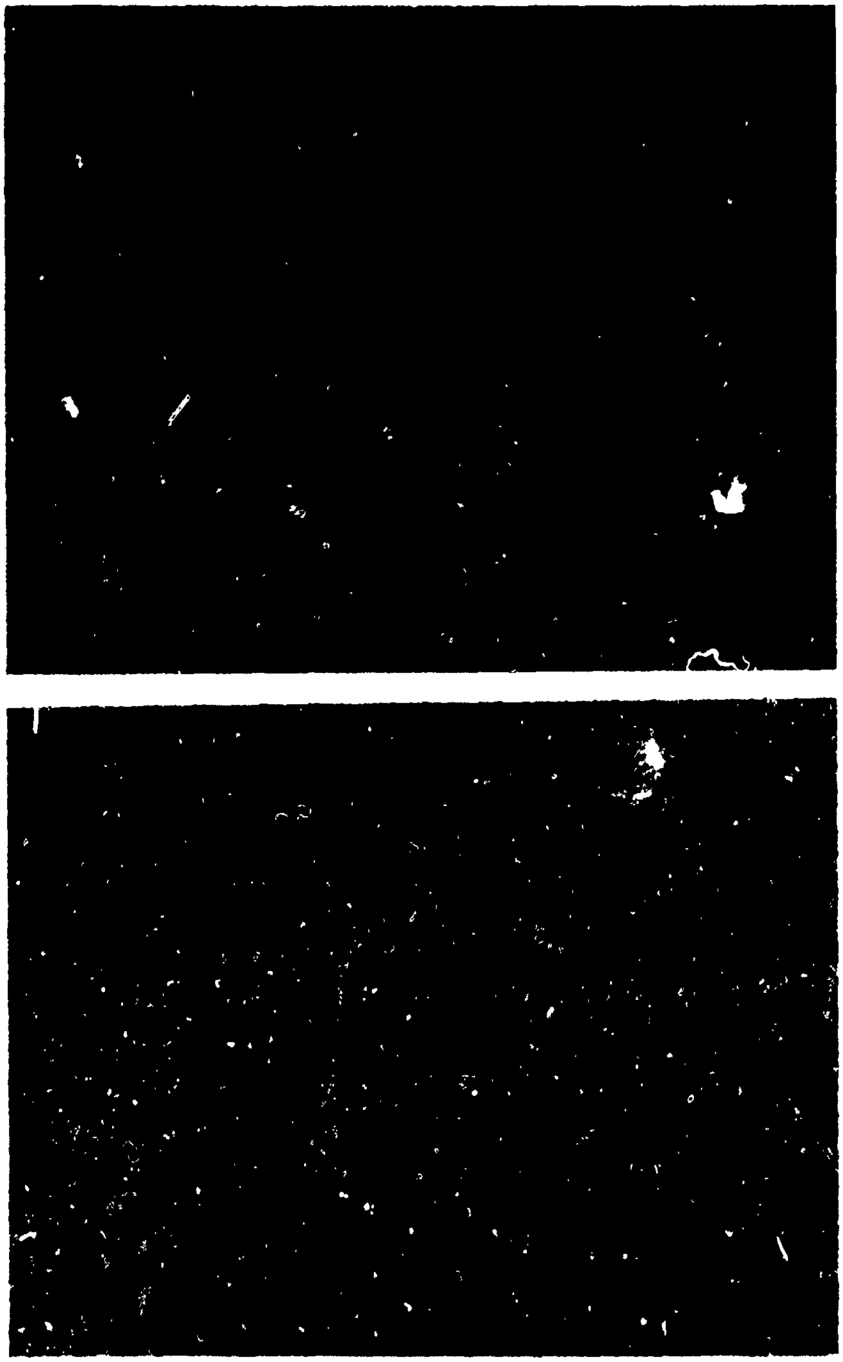

FIGURE 31 - REPLICA MICROGRAPHS OF MILL ANNEALED CONDITION ASTROLOY BASE METAL. X3,700 (TOP), X13,000 (BOTTOM)

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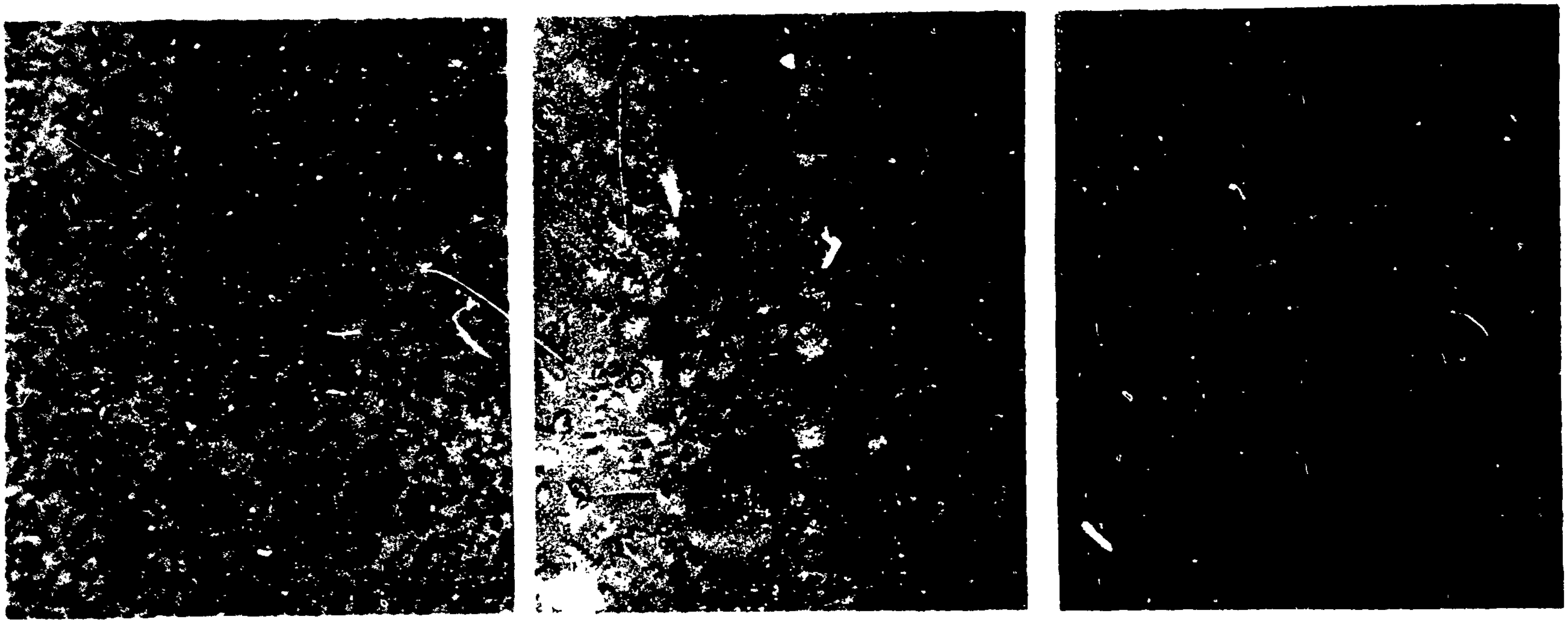

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FIGURE 32 - REPLICA MICROGRAPHS OF ASTROLOY SHEET IN THE "1975°F-16 HRS (WQ) + 1850°F-24 HRS (WQ)" PRE-WELD CONDITION. X1000 (LEFT), X2500 (CENTER AND RIGHT).

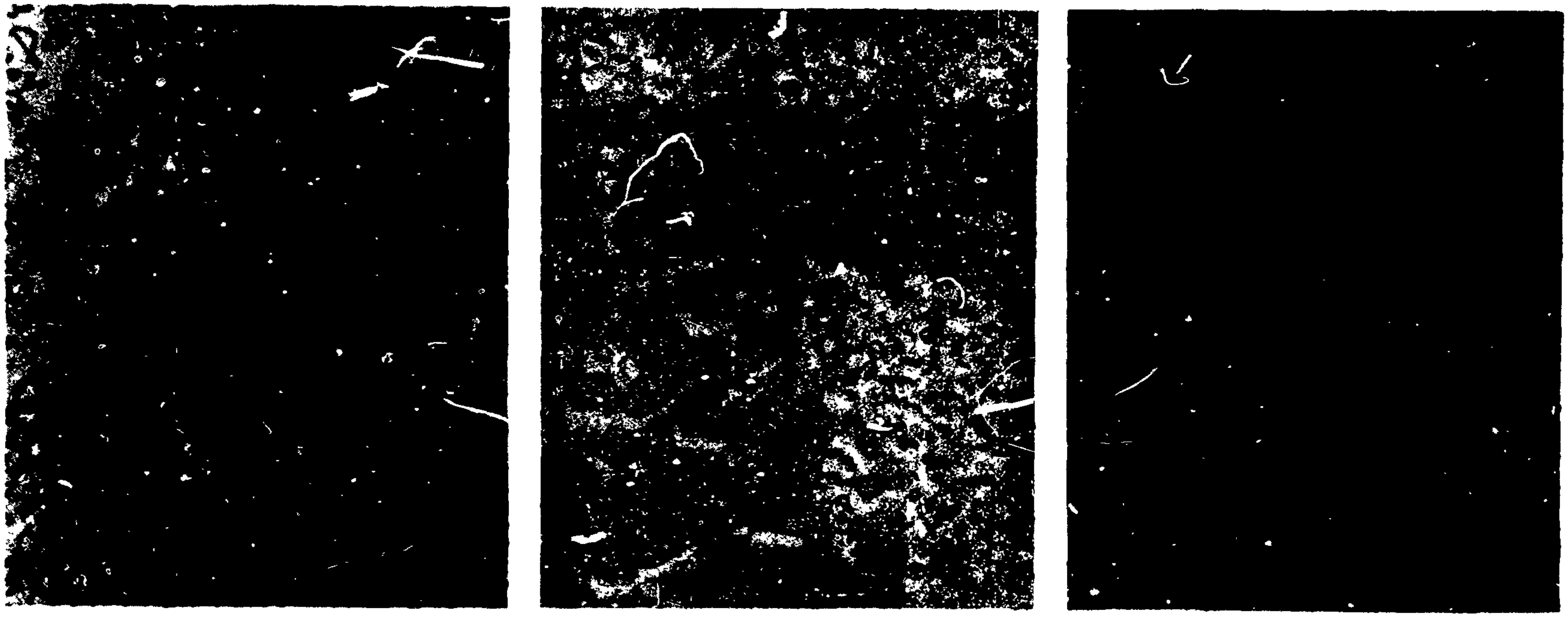

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FIGURE 33 - REPLICA MICROGRAPHS OF ASTROLOY SHEET IN THE "1975°F-16 HRS (FC TO 1850°F-24 HRS (WQ)" PRE-WELD CONDITION. X1000 (LEFT), X2500 (CENTER AND RIGHT)

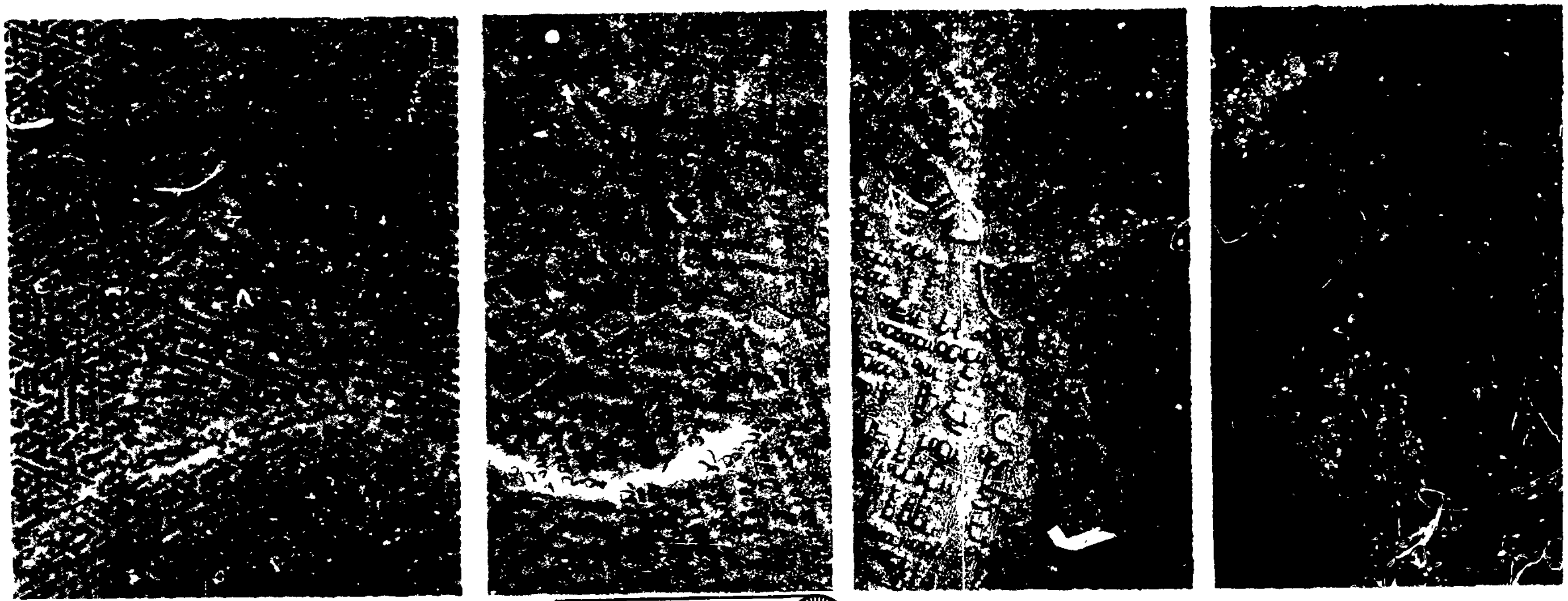

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FIGURE 34 - REPLICA MICROGRAPHS OF ASTROLOY SHEET IN THE 2140°F-4 HRS (FORCED AIR COOLED) + 1975°F-16 HRS (COOL 100°F/HR TO 1850°F-4 HRS (COOL 50°F/HR TO 1650°F, 100°F/HR TO 1050°F)" PRE-WELD CONDITION. X1000 (LEFT), X250 (REMAINING THREE).

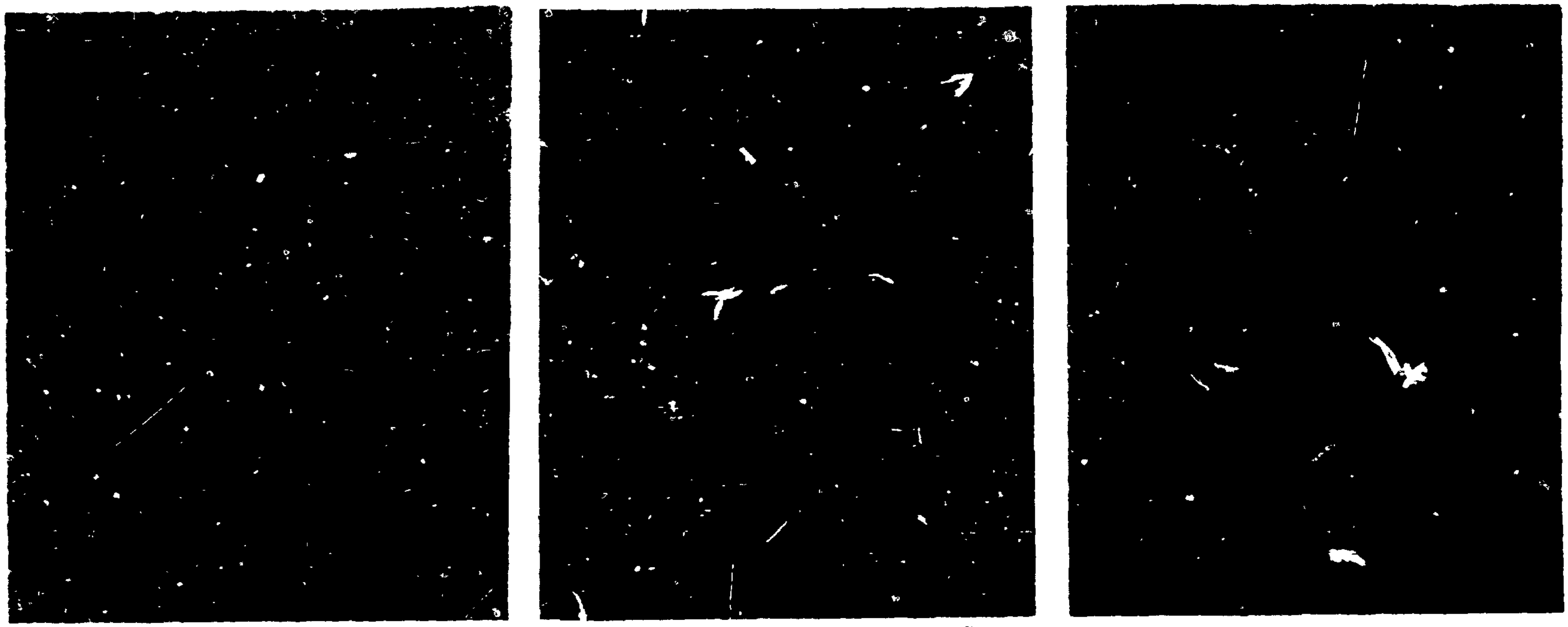

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FIGURE 35 - REPLICA MICROGRAPHS OF ASTROLOY SHEET IN THE "2125°F-16 HRS (FC TO 1850°F) + 1850°F-4 HRS (WQ) + ON-HEATED TO 1000°F" BASE METAL CONDITION. X1000 (LEFT), X2000 (CENTER AND RIGHT)

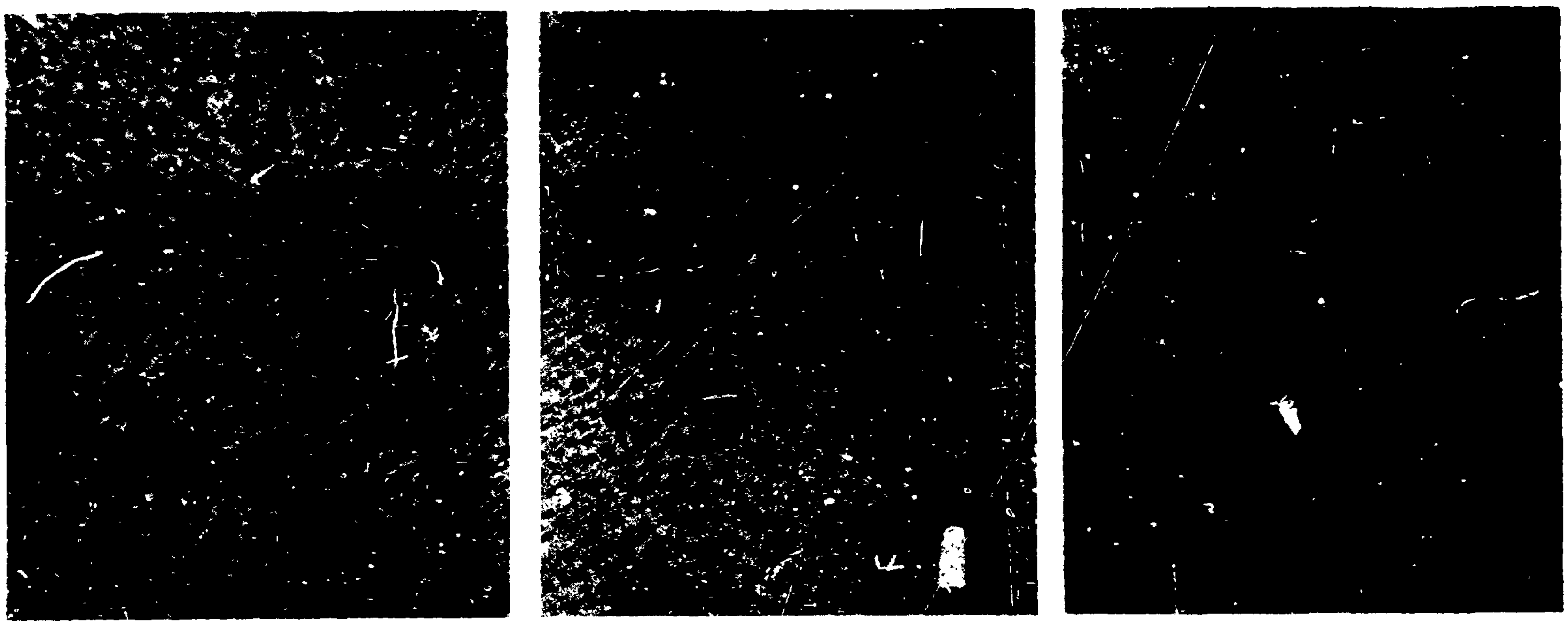

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FIGURE 36 - REPLICA MICROGRAPHS OF ASTROLOY SHEET IN THE "2125°F-16 HRS (FC TO 1850°F) + 1850°F-4 HRS (WQ) + ON-HEATED TO 1600°F" BASE METAL CONDITION. X1000 (LEFT), X2000 (CENTER AND RIGHT).

The typical mill annealed condition microstructure is shown in the replica micrographs of Fig. 31 for reference purposes. The structure contains voluminous fine γ' precipitate with some coarse γ' particles; the grain is not apparent in these micrographs, however, it is quite fine (about ASTM Grain Size 10). The replica micrographs of Figs. 32 and 33 are for the "1975°F-16 hrs (WQ) + 1850°F-24 hrs (WQ)" and the "1975°F-16 hrs (FC to 1850°F) + 1850°F-24 hrs (WQ)" pre-weld heat treatments, respectively. The basic difference between these treatments is in post-solution cooling; the first is water quenched and the second is furnace cooled to 1850°F; hardness data for the water quenched condition may be found in Table XII and Fig. 27. Both fine and coarse γ' are found with each condition; the fine γ' in these structures is considerably coarser than that found with the mill annealed condition. The microstructure differences associated with these conditions are consistent with the variations in heat treatment. The water quenched condition (Fig. 32) exhibits finer and more numerous smaller γ' particles and fewer and less coarse larger γ' particles than does the furnace cooled condition (Fig. 33). The finer and more numerous smaller γ' particles seen in Fig. 32 are due to precipitation events occurring during on-heating to 1850°F, whereas the fewer and slightly more coarse γ' particles seen in Fig. 3 may be attributed to precipitation events at higher temperatures.

The rather decorative replica micrographs of Fig. 34 are for the P & W developed pre-weld overaging treatment; on-heating hardness data for this condition may be found in Fig. 28 and results of crack-susceptibility tests in Table XIV. The microstructure consists of small γ' particles throughout the matrix with larger particles located along many of the grain boundaries. Many of the grain boundary regions also exhibit precipitate free zones as seen in Fig. 34. These regions would most likely be sites of localized yielding and microplasticity effects under high restraint conditions. The microstructures of Astroloy given the "1975°F-16 hrs (FC to 1850°F) plus 1850°F-4 hr (WQ)" pre-weld heat treatment noted in Fig. 28 were quite similar to those observed with the P & W developed heat treatment (see Fig. 34). In order to determine the on-heating precipitation behavior of the "1975°F-1850°F" overaged condition just described, specimens were on-heated to 1000°F and to 1600°F and then prepared for replica microscopy. The precipitate-free zones are still apparent after heating to

1000°F as shown in Fig. 35, whereas the zones do not exist upon heating to 1600°F. This is somewhat consistent with the on-heating hardening data shown in Fig. 28 which indicates a slight metastability in the 1400°-1600°F temperature range. Considering the failure of crack-susceptibility specimens during welding noted in Table XIV, it might be conjectured that these precipitate-free zones do not exist in the immediate (high temperature) region of the heat-affected zone and may influence (or increase) the susceptibility of this area to cracking during welding. This effect would obviously be compounded by the fact that the microstructure is now coarse grained and located in a high stress concentration field (at the toe of the weld). Thus the stress relaxation capability of this region is markedly reduced and the resistance to heat-treat and/or weld cracking is lowered.

SECTION V

CONCLUSIONS

The resistance to post-weld heat-treat cracking of Rene' 41 was found to be influenced by the specific pre-weld base metal solutioning time and temperature used and strongly dependent upon the effectiveness of post-solution quenching. The susceptibility to cracking was shown to increase with increasing solutioning temperature and/or time (above 1975°F) and decreasing post-solution quench rate. On-heating hardness data obtained for various heats of Rene' 41 and one heat of Waspaloy and Astroloy illustrated the "slack-hardening" effect which shows how the bulk base metal microstructure can plastically accommodate (stress relieve) existing restraint stresses thus avoiding on-heating cracking. Fortuitously, the 1975°F-1/2 hr (WQ) pre-weld solutioning treatment appears to be an optimized solution treatment for the intermediate hardner content alloys such as Rene' 41 and Waspaloy from both a standpoint of avoiding heat-treat cracking and pre-weld processing economics.

On-heating Rene' 41 restraint test data showed that a small amount of pre-weld cold work (2% in this case) did not increase the susceptibility of the pre-weld solutioned (1975°F-1/2 hr, WQ) condition to heat-treat cracking. Previous work(1) showed that 5% cold work was detrimental. This is an important finding in that sheet forming operations in which small amounts of cold work are typically encountered can be performed in the "soft" condition without decreasing the base metal's resistance to heat-treat cracking. Crack-susceptibility results showed that the effects of oxidation due to inadequate shielding during welding did not increase the susceptibility of Rene' 41 to heat-treat cracking. This is an aspect of fabrication, especially when tack welding, which could be of concern from the standpoint of the effects of prior oxidation surface damage on heat-treat cracking behavior.

A slightly modified circular-patch type on-heating specimen was developed and evaluated to establish the effects of furnace atmosphere on heat-treat cracking. This specimen was essentially an integral-retort which permitted the correlation of the

effects of an air atmosphere and a vacuum environment on cracking behavior. Restraint test results showed that the use of vacuum increased the "gross cracking temperature" about the same order of magnitude as the use of lower weld energy input techniques reported earlier(1). This increase in resistance to heat-treat cracking obtained by eliminating oxygen is not as marked an effect as reported by prior investigators. One of the major effects noted with the vacuum test series was the improved resistance to rapid crack extension; acoustic emission analyses revealed that the rate of crack growth was lower with the vacuum tests than with the air tests. It should be noted that the difference in heat-treat cracking characteristics in vacuum versus air may be more pronounced with restraint stress conditions less severe than those produced with the test specimen used for these studies.

Isothermal crack-susceptibility data indicated that pre-weld solutioned Rene' 41 sheet converted from REP (Rotating Electrode Process) powder bar stock is more susceptible to heat-treat cracking than pre-weld solutioned commercial sheet product. Metallographic studies showed that heat-treat cracking experienced by the REP material was associated with porosity at the fusion-heat affected zone interface, an effect not uncommon with the welding of powder products. The nature of cracking observed with the REP sheet specimens indicates that the material has considerable inherent resistance to cracking. This is suggested by the fact that through-the-thickness cracking did not occur and the part-through cracks were "blunted" by in-plane micro-delaminations.

Procedure verifications in the form of on-heating crack-susceptibility tests and on-heating hardening responses have shown that the 1975°F-1/2 hr (WQ) is effective for avoiding heat-treat cracking in both Rene' 41 and Waspaloy, including three heats of Rene' 41 and one heat of Waspaloy.

On-heating crack-susceptibility test results showed that increasing solutioning temperature and/or time at temperature increased the susceptibility of Astroloy to cracking during welding. Overaging treatments, which appeared favorable from the standpoint of producing low pre-weld hardnesses and sluggish responses during on-heating, were

found to be detrimental; these treatments were found to markedly increase the susceptibility of Astroloy to cracking during weld cooling. The most likely pre-weld heat treatment approaches for avoiding cracking in the high hardener content alloy such as Astroloy appear to involve duplex heat treatments combining solutioning and γ' precipitate conditioning thermal cycles. The data obtained for Astroloy would suggest that grain growth, due to prolonged high temperature exposure, is more detrimental with respect to heat-treat cracking in the high hardener content alloys than in the intermediate hardener content compositions.

SECTION VI
RECOMMENDATIONS FOR FUTURE WORK

(1) Investigate new pre-weld heat treatments for the higher hardener content alloys, such as Udimet 700, for avoiding post-weld heat-treat cracking. These should be combined solutioning-precipitate conditioning heat treatments directed specifically to manipulating γ' to obtain desirable on-heating structural characteristics.

(2) Investigate the use of pre-weld heat treatments for decreasing the susceptibility to weld and/or heat-treat cracking of presently used nickel-base superalloy powder metallurgy products such as sheet, bar, and forgings. Emphasis to be placed on method of powder production and consolidation and the development of heat-treat repair welding procedures for such commercial components as engine blades.

(3) Establish specification requirements for pre-weld heat-treat processing of the superalloys for use in welded fabrication. Support data should include microstructure characterizations as well as results from crack-susceptibility and mechanical tests representing an adequate statistical sampling of the pertinent parameters.

(4) Continue metallurgical and cracking studies with the nickel-base superalloys designed towards developing a model for heat-treat cracking. To date, the specific mechanism(s) involved in post-weld heat-treat cracking has been speculated with little, if any, experimental data support.

(5) Continue crack-susceptibility studies directed towards defining more quantitatively the specific influence of restraint stresses on post-weld heat-treat cracking. In the past efforts, data were obtained with a specific state of residual stress with emphasis on the susceptibility to heat-treat cracking during the on-heating interval. It is suggested that future efforts be directed towards lower restraint condition tests and isothermal cracking behavior.

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	Air Force Materials Laboratory Air Force Systems Command Wright-Patterson Air Force Base, Ohio

13 ABSTRACT

This is a report of a program concerned with the development of procedures for avoiding post-weld heat-treat cracking in nickel-base superalloy weldments. Crack-susceptibility test procedures and an acoustic emission technique developed in a prior program were used in this program. Crack-susceptibility data were obtained for a number of studies including the effects of varying pre-weld solutioning treatments, effects of pre-weld cold work, effects of oxidation damage due to inadequate shielding during welding, evaluation of Rene' 41 powder metallurgy sheet material, procedure verifications, and the evaluation of varying pre-weld base metal heat treatments for avoiding heat-treat cracking in Astroloy. The 1975°F-1/2 hr (WQ) solutioning treatment was found to be an optimized pre-weld heat treatment for avoiding post-weld cracking in Rene' 41. Post solution quench rate was found to be a critical variable with slower quenches increasing the susceptibility of Rene' 41 to heat-treat cracking. Small amounts (to 2%) of pre-weld cold work did not reduce the effectiveness of pre-weld solutioning treatments for avoiding heat-treat cracking. The 1975°F-1/2 hr (WQ) pre-weld solution treatment was found to be effective in avoiding post-weld heat treat cracking in three heats of Rene' 41 and one heat of Waspaloy. Promising pre-weld overaging treatments for Astroloy were found to be detrimental in that they markedly increased the material's susceptibility to cracking during welding. The report also covers on-heating hardening response and detailed microstructure studies associated with varying pre-weld base metal conditions.

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Security Classification

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	ROLE	WT	ROLE	WT	ROLE	WT
nickel-base superalloys						
Rene' 41						
Astroloy						
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gas tungsten-arc welding						
post-weld heat-treat cracking						
crack-susceptibility testing						
acoustic emission testing						
acoustic characterization of cracking						
thermal-mechanical treatment						
mechanical properties						
metallography						